A STUDY OF HIGH SPEED STEEL

THESIS FOR THE DEGREE OF M. S.
Edwin A. Brophy
1933

THESIS

Steel

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Henry E. Pueblan

Some Observations Of The Microscopic And Physical Changes Resulting From Heat-Treatment

And

An Investigation Of The Dilatometric And Thermal Changes Resulting from Heat-Treatment

Of

High Speed Steel

Thesis

Submitted To The Faculty
Of
Michigan State College

In Partial Fulfillment

Of The

Requirements For The Degree

Of

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Metallurgy

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M S

THESIS

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The writer wishes to express his gratitude and thanks to Prof. H. E. Publow for his unending kindness and many helpful thoughts and suggestions.

The writer also expresses indebtedness to Mr. N. I. Stotz for samples and information of details of practice this heat of steel underwent.

ABSTRACT.

This thesis, a study of high speed steel, consists of a short historical sketch, the manufacture from the raw products going into the electric furnace, the chemistry of melting and the fabrication through to the final annealing operation.

A comprehensive metallographic study is shown with photomicrographs of quenching temperatures from 1400° to 2400° F. A similar study is shown of various drawing or tempering temperatures; this includes the recommended hardening practice.

A microscopic study shows results obtained using several etching reagents.

A very complete dilatometric study with curves from the Chevenard Thermal Analyzer is given.

HISTORICAL

Before entering into a detailed study of this analysis as used today, it might be well to consider the history of our present day high speed steel.

Originally all tools were made from a good grade of high carbon tool steel. Inasmuch as the design of machine tools was not up to present requirements of feed and speed, these tools served their purpose in a manner. However as the speed of shop machines were increased, a deep need was felt for greater efficiency and performance in tools. The result was a tool steel developed by Mushet in England and generally referred to as Mushet's Air Hardening Steel. This was a high manganese (austenitic) type of steel, and received its name from the property of hardening when cooled in air. This air-hardening property was due to its manganese content.

With the end of the nineteenth century and the beginning of shop efficiency and big production, it was imperative that tool costs be cut down; also tools that could stand the abuse of the improved machine tools. Much research brought forth the epochal discovery at the beginning of this century of our present 18: 4: 1 type of high speed steel by Taylor and White of Bethlehem Steel Company.

It is probably the greatest single advancement ever made in tool steels.

The treatment which they recommended for the standard type of high speed steel has changed but little since that time.

During the last few years special high speed steels containing high percentages of carbon, chromium, vanadium, molybdenum, cobalt, etc; of varying proportions have been developed. The treatment for these steels will vary somewhat from that recommended for the standard type.

The cost of these steels is prohibitive for general utility purposes, and they only find application where the cost of tools is secondary.

MANUFACTURE

This study was made with samples kindly supplied by the Braeburn Alloy Steel Corporation, and according to their analysis is as follows:

Carbon ---------------.69
Manganese-------------.18
Phosphorus------------.014
Sulphur---------------.011
Silican---------------.29
Chromium-------------4.09
Vanadium-------------1.07
Tungsten------------17.65

The above is a standard type of what is generally referred to as the 18: 4: 1 type of high speed steel and except for the higher carbon content varies little from the original analysis as brought out by Taylor and White in their epochal discovery at the beginning of this century.

Commercial high speed steel is usually made by the electric furnace process. The Heroult arc-type is most commonly used. Some of it is made in the electric induction furnaces on a small scale.

This particular steel was made in a six ton Heroult arc furnace. Figure 1 gives the actual history of the heat of

Figure I.

Melting Log.

2M-6-31-JLH

ELECTRIC FURNACE DEPARTMENT

HEAT NO. 2742 DATE 5·12·32 GRADE V·H·S SIZE OF MOULDS 12" 8" HEATS ON ROOF 43 HEATS ON LINING 43 HEATS ON BOTTOM 43 NOZZLE 1¼"

METALLIC CHARGE

MAT'L	WEIGHT	TIME ADDED
H·S· Scrap	3500	
Tools (H·S·)	4500	
Iron	1500	
Nails	1000	
Scheelite	1300	
WASHED METAL		
FE.CHROME	250 (low Carbon)	
FE. MN.	–	
FE. SIL.	30	
FE. VA.	160	
FE. W.	225	
Zr Silicide	23	
TOTAL		

NON-METALLIC CHARGE

MATERIAL	WEIGHT
LIME	275
FLUORSPAR	300
COKE	
SAND	
SYNDOLAG	
MAGDOLITE	350
LIMESTONE	
Ca C₂	400
TOTAL	

DELAYS

CAUSE	HRS.	MIN.
CRANE		
ADDING ELECTRODES		
LABORATORY		
MOULDS		

REMARKS

CONDITION OF BOTTOM Slightly cold from standing, but O.K.

CONDITION OF LADLE Very good

CONDITION OF MOULDS All smoked with resin. O.K

TEMPERATURE A trifle on the cold side.

MELTER

TIME OF HEAT

TAPPED PREVIOUS HEAT		A. M. P. M.
MAKING BOTTOM		
CHARGING	4:00	
POWER ON	4:20	A. M. P. M.
MELTING DOWN	6:20	
DEOXIDIZING	2 hrs 55 mins	

TAPPED	9.15	~~A.M.~~ P. M.
POWER ON HEAT		
TOTAL TIME ON HEAT		
METER AFTER	411000	
METER BEFORE	406500	
K. W. H.	4500	

PRODUCTION

NO. OF INGOTS	SIZE	WEIGHT
2 Solid	12"	4390
7 "	8"	5600
½ "	8"	405

SCRAP

DESCRIPTION	WEIGHT
BUTTS	400
LADLE SKULL	400
PITT	
HOT TOPS	

	C.	MN.	PHOS.	SUL.	SI.	CR.	VA.	TUNG.	NI.	CO.		TIME TAKEN	TIME REPORTED
ORDERED	68/72	15/30			15/30	3.75/4.25	.85/1.00	17.50/18.50					
1ST PRELIM	·63					2.67	·46	17.15					
2ND PRELIM	·67												
3RD PRELIM													
4TH PRELIM													
FINAL	·69	·18	·014	·011	·29	4.09	1.07	17.65					

steel from the raw materials to the steel in the ladle and ingots.

At this point I think it might be well to make brief mention of the chemistry of the electric arc melting process for steel manufacture. The standard practice, as used in this case, is to charge the scrap high speed steel, nails, virgin iron and necessary metallic alloys into the empty furnace with sufficient lime to form a protective coating on melting. The current - low voltage and high amperage - is immediately turned on.

This charge was sufficiently melted in two hours to start adding flourspar to make lime slag less viscous. Where the charge contains such impurities as phosphorus, sulphur, and manganese, it is necessary to first run an oxidizing slag, sometimes called the dephosphorizing period. This period consists of the removal of Mn, Si, and P, at low temperatures by the following reactions:

$$Mn + FeO = MnO + Fe$$

The MnO rises and dissolves in the slag which is later deoxidized by the carbon.

$$MnO + C = CO + Mn$$

The Mn returns to the bath and the Co is liberated.

$$Si + 2\ FeO = SiO_2 + 2\ Fe$$

The SiO_2 is fluxed by the lime slag and the Fe returns to the bath.

$$3CaO + P_2O_5 = (CaO)_3 \cdot P_2O_5$$

The calcium phosphate goes to the slag and is removed when slag is scraped off before finishing or reducing slag is formed. If impurities are not great the oxidizing slag is not removed, but made strongly reducing for finishing by carbon additions in the form of graphite, or charcoal or ground coke. During the oxidizing period a small amount of sulphur is reduced from the metal:

$$FeS + CaO + C - Fe + CaS + CO.$$

The ideal change for high speed steel should be so proportioned as to require little or no oxidizing. Thus with a carefully selected charge having a low percentage of non-metallic impurities one slag only is required. This eliminates the necessity of removing the first or oxidizing slag and exposes the molten metal to rapid oxidation by air. Using the one slag throughout the entire process necessitates careful handling and constant "doctoring" with lime, spar, and coke to bring it from a slightly oxidizing condition to a strongly reducing one. The final slag, to ensure complete deoxidation, should slack from the presence of calcium carbide.

If the bath of metal has been well rabbled the presence of CaC_2 should indicate absence of oxides:

$$CaC + 3FeO - 3Fe + CaO + 2CO$$

The CaC_2 being reacted upon by FeO, or other oxides, cannot long exist.

By reference to figure 1 it will be noticed that an addition of 225 pounds of ferro-tungsten was necessary to finish with the correct analysis. Large additions of heavy metallics such as ferro-tungsten while finishing is not considered the best practice. Ferro-tungsten is heavy and difficult to get into solution. Steady rabbling at such a time is necessary, otherwise the finished product shows tungsten segregation and considerable heterogenity. Therefore great care should be exercised in the original calculations to melt down as nearly as possible to the desired. final analysis. It should be noted that an addition of 25 pounds of zirconium silicide was added to the ladle on tapping for degasifying purposes.

POURING INGOTS AND FABRICATION

This heat of steel was tapped in a slightly cold condition (see Figure 1) and therefore no time was lost in teeming into ingots which had been heated and thoroughly smoked with resin fumes.

The ingots cast were 12" square and 8" square with the big end up and a heavy heated clay sinkhead to act as a feeder to the ingot proper. This concentrates the shrinkage ("pipe") and permits much less cropping with a sounder and safer ingot.

The ingots on being "stripped"were - as is the practice at this mill- charged before they became cold into a furnace about 1600°F. When thoroughly heated to an approximate temperature of 2100°F, they were forged by the regular mill practice for a steel of this type, namely, reduction under hammer to one half of their original cross section or one fourth of their area. That is the 12" ingots were cogged to 6" square and the 8" ingots were cogged to 4" squares. The surface of these billets were ground all over while they were still warm, approximately 400°F. It is the practice at some mills to anneal billets before grinding. This assists in locating small surface "checks" and "seams" as well as adding greater plasticity to the steel, thus reducing the risk of setting up grinding cracks from induced heat and strains. It also necessitates somewhat less care in the grinding operation. The samples used in this investigation.

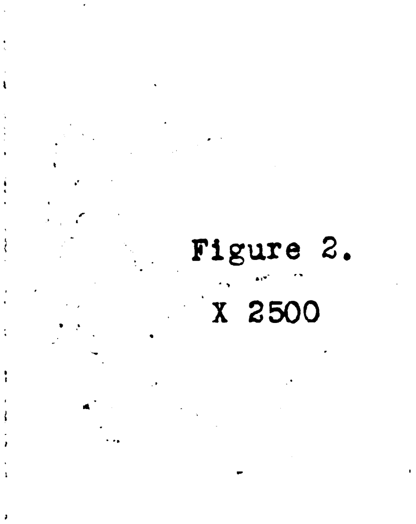


Figure 2.

X 2500

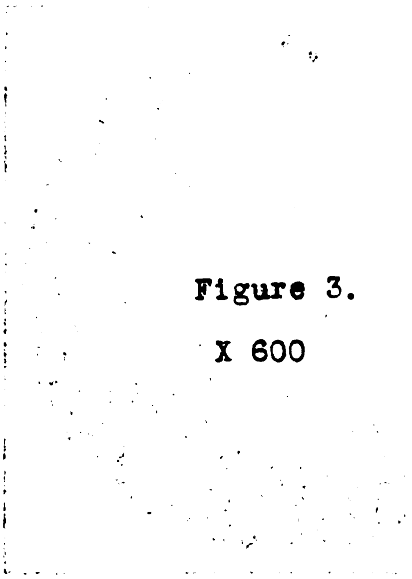
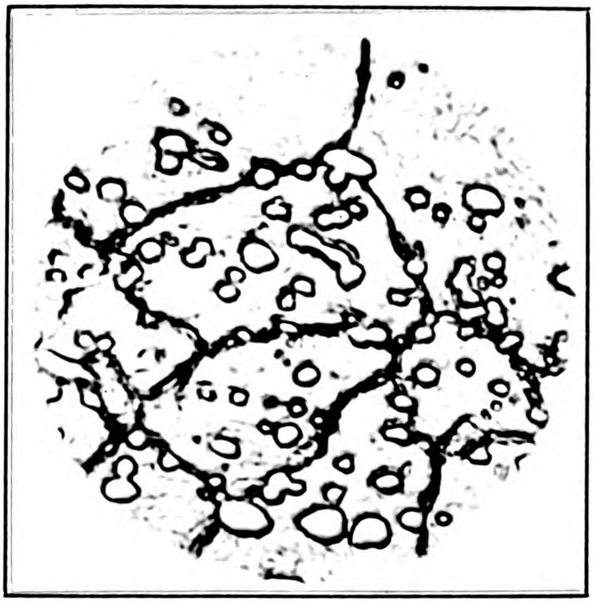
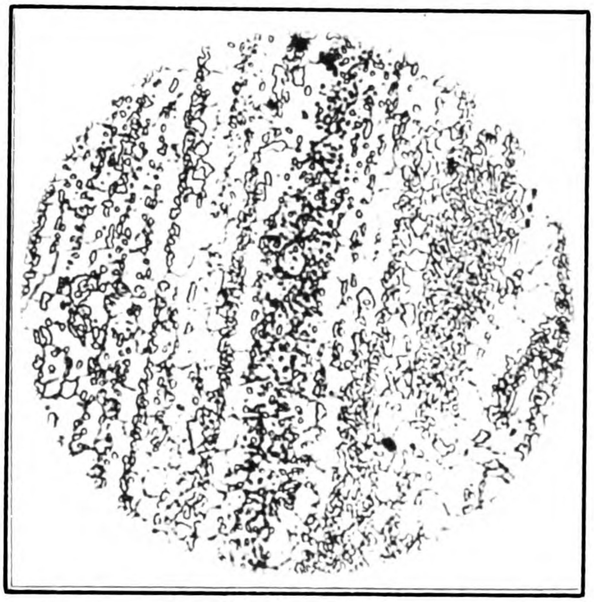


Figure 3.

X 600





came from an 8"square ingot which was cogged to a 4" square, and then finished on a 10" rolling mill to 15/32" by 21/32".

Figures 2 and 3 show the large carbide grains which have probably broken down in working and heating from the eutectics formed during the solidification of the ingot. This particular specimen was taken from a 1 1/2" square section before annealing The lower magnification clearly defines the parallel banding and flow of metal in the direction of rolling. The higher magnification shows definite evidence of a structure beneath the etched surface, that is, a superimposed structure. The small emulsified carbides are probably precipitated as such. The repeated heatings for forging and rolling dissolves much of the carbide particles and they are again precipitated. But such action involves chiefly the smallest particles, which present the greatest surface per unit volume.

According to Grossman and Bain the relative attack upon the larger carbide particles is very slight, and they assume that the obliteration of concentrated carbide regions is largely mechanical. Thus after the final working, it is found that we have two distinct sets of carbide particles: The small spherodized carbides precipitated from solution and the larger ones (usually banded longitudinally to working) from the mechanical disintegration of the coarse eutectics. See figure 4 which shows this condition very clearly.

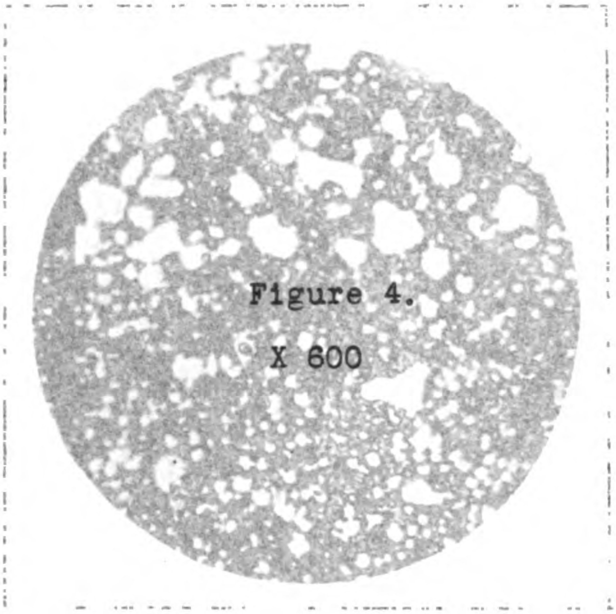


Figure 4.

X 600

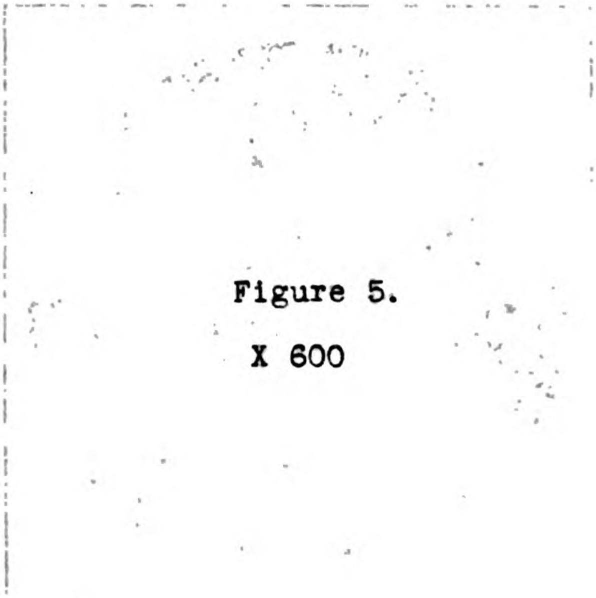


Figure 5.

X 600

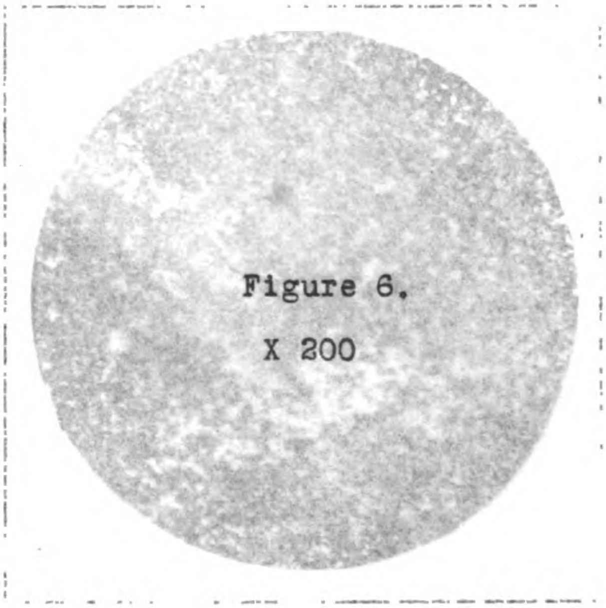
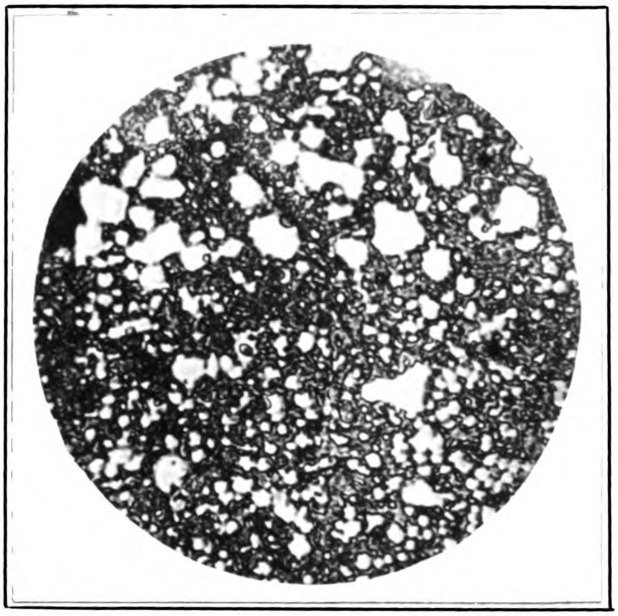
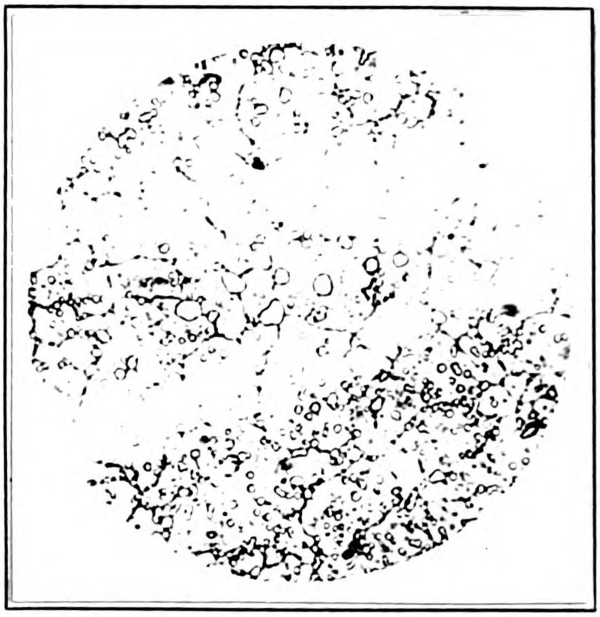


Figure 6.

X 200





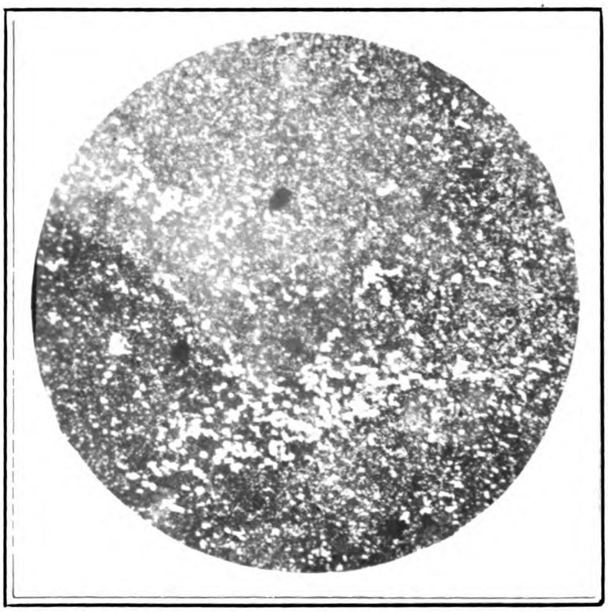


Figure 5 shows what is probably a richly alloyed band of carbides. This runs in several bands through the central portion of the bar, and is not an uncommon condition in high speed steel. This specimen was hardened to give the banding greater prominence when etched.

The reheatings apparently had little or no effect on this cored condition. The partial breaking down was by mechanical work and flowed in the direction of the work. Good examples depicting lack of homogenity of well worked high speed steel will be found in Figures 4 and 6. One of the characteristics. of this analysis is the inability of its component parts to diffuse, even after long anneals and repeated heatings. These samples show the streaky carbide region wherein the carbide particles are dense and plentiful. This perhaps indicates an inequality of the chemical composition. This difference in composition could easily be due to the difference in composition of the first and last (eutectic) grains to be formed. The differential is seldom equalized by diffusion - the steel being too sluggish.

It has been suggested that the red-hardness of high speed steel might be partly due to this atomic immobility or sluggishness resulting from inhomogenity.

According to mill reports, this particular heat of steel gave an excellent recovery and showed no indication of trouble.

ANNEALING

High speed steel after finishing on the rolls or hammer shows considerable hardness from air cooling. In order to put it in a state where it can be readily machined, cropped, and inspected it is necessary to thoroughly anneal.

The Brinell hardness before annealing is quite variable, but is usually well above the 500 number. After annealing it varies according to the condition of anneal and hardness desired. It is usually in a range between 196 and 248 B.H.N. An extremely low Brinell hardness is seldom desired as it seems to be mushy and tears in machining. For general purposes a B.H.N. of 228 is recommended.

Figure 7 and 8 show the microstructure of this steel well annealed. It will be observed that the carbides are small and fairly well dispersed. In order to get the maximum machine ability, it is necessary to remove all traces of austenite and martensite which may be present from the mill operations. (See Figure 2)

The small precipated carbide particles shown in Figure 5 offer little hardening effect, and therefore lend to machinability.

The generally accepted annealing practice in tool steel mills

Figure 7.

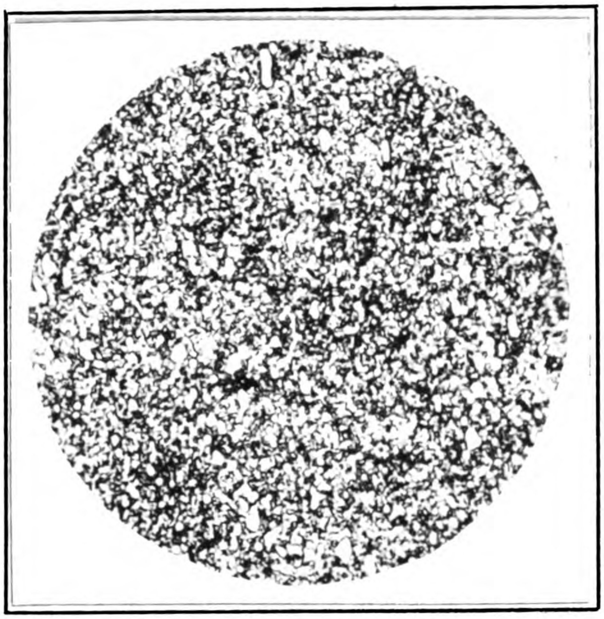
X 600

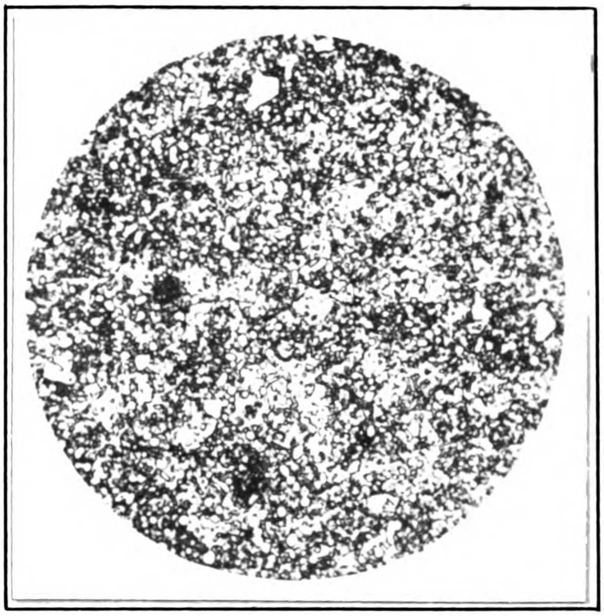
Figure 8.

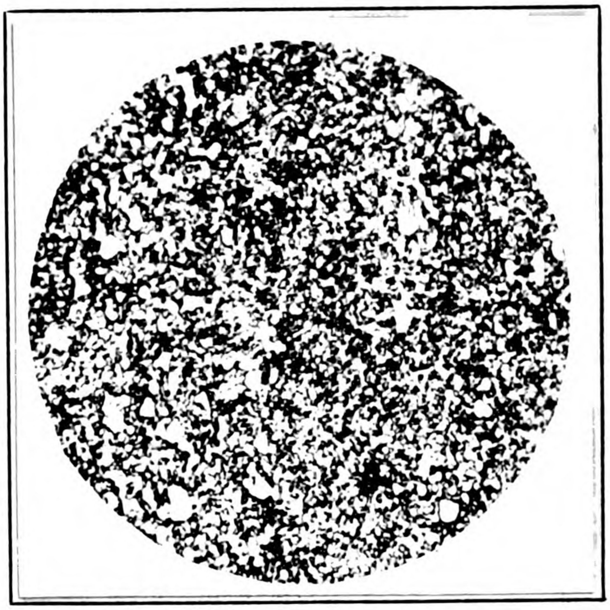
X 600

Figure 9.

X 600







for steel of this type is to anneal in steel or iron pipes with or without a packing material. This steel was pipe annealed and packed with 80% washed river sand and 20% wood charcoal. As the charcoal becomes "spent" it is constantly enriched in order to keep the carbon content between the limits of 15% and 25%. With such a mixture the characteristic soft or decarburized skin resulting from annealing is prevented. Thus, a superior hardening surface is obtained. The loaded pipes are charged into a cold or warm furnace and slowly brought up to approximately 1550° F. It requires 7 or 8 hours to reach this temperature, and it is then held at this point for about 6 to 8 hours. Gas or electrical current is then reduced and finally cut off. The time of cooling is about 6 hours. The above times are approximate and vary with the nature and mass of the charge. When the furnace has cooled to about 800°F. the pipes can be removed from the furnace, but the steel is left in the pipes until nearly cool enough to handle. The speed or rate of cooling below 1000°F. is not considered important.

The series of micro structures as shown in Figures 7,8,9,10 and 11 give the structure existing after a thorough annealing. Both cross and longitudinal sections are shown. Figure 7 is a longitudinal section and shows a well dispersed carbide condition. This section was taken from a ½" round bar. The worst spot was selected for the photomicrograph.

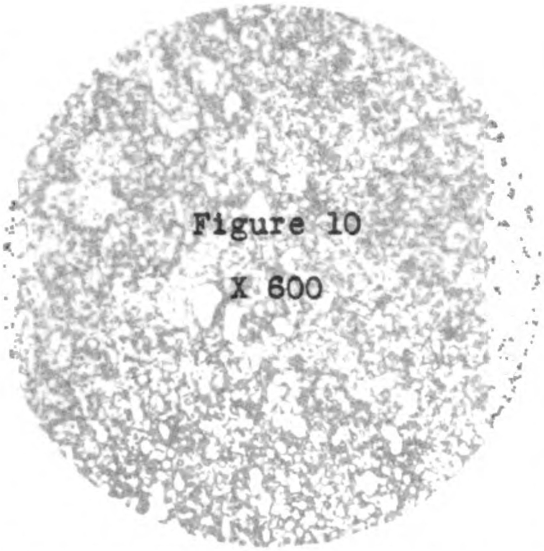


Figure 10

X 600

Figure 11

X 600

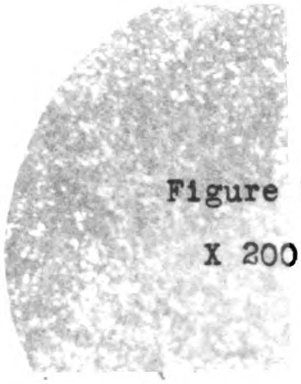
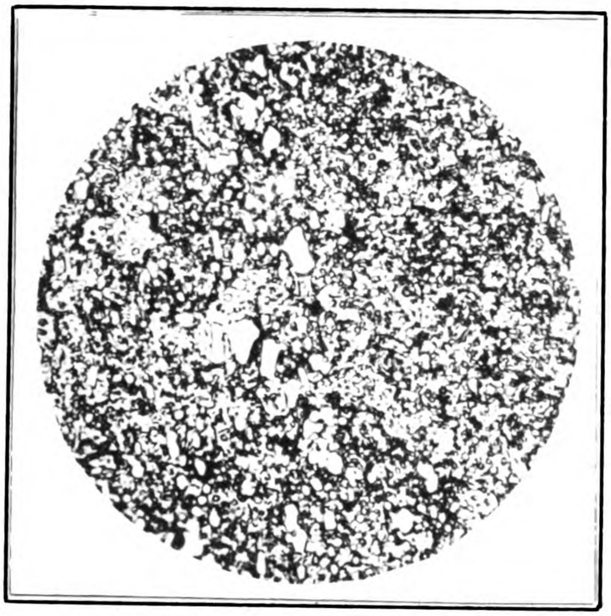
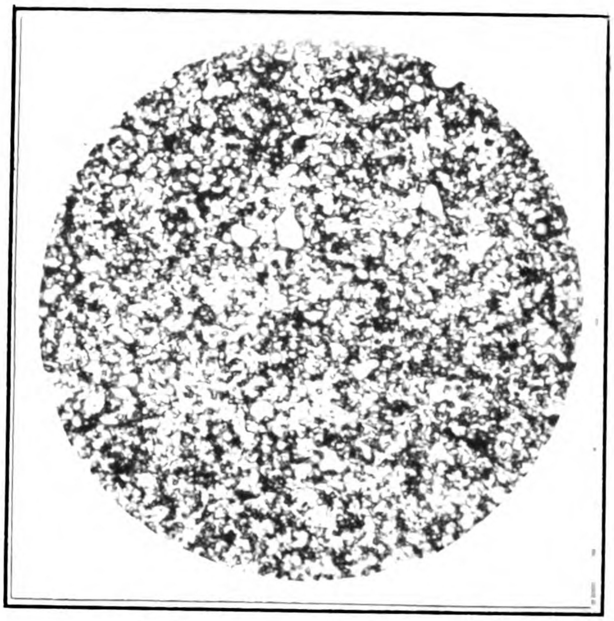
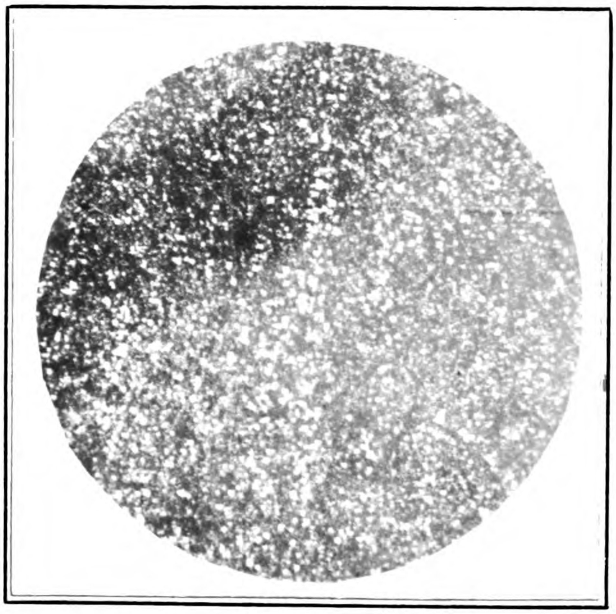


Figure 12

X 200







In Figure 8, which is taken in a longitudinal direction, there is just a faint trace of banding of the primary carbides in the direction of rolling. No matter how much reduction is given to the steel this is a very difficult condition to destroy entirely. Further rolling merely means further elongation and streaks resembling ghosts lines result.

Figure 9 is the same ½" diameter section shown in Figure 7, except it was taken from a cross section of the specimen. The worst spot was again selected for the photomicrograph. So it demonstrates clearly how well the steel has been worked and annealed. The carbide distribution is exceptionally good.

Figure 12 was taken at a lower magnification in order to emphasize the banding or segregation of carbides. This is a longitudinal section and on close observation it is possible to discern the crystal (grain) arrangement of the carbides in the direction of rolling. It is however very slight. It will be observed, series 13, 14, and 15, that there two distinct sets of carbide grains, that is, grains of two sizes. The large grains may be regarded as primary carbides existing from the original ingot or eutectic structure. The smaller carbide grains have resulted from eutectic disintegration and precipitation on reheating for working.

The photomicrographs in Figures 16 and 17 show a longitudinal and cross section respectively, in the unetched condition

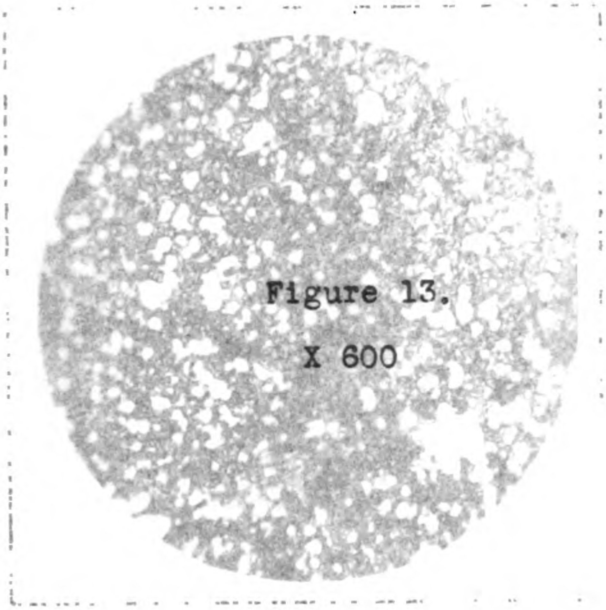


Figure 13.

X 600

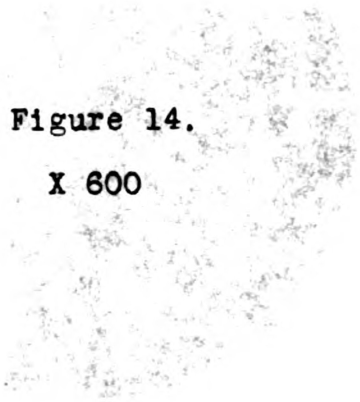


Figure 14.

X 600

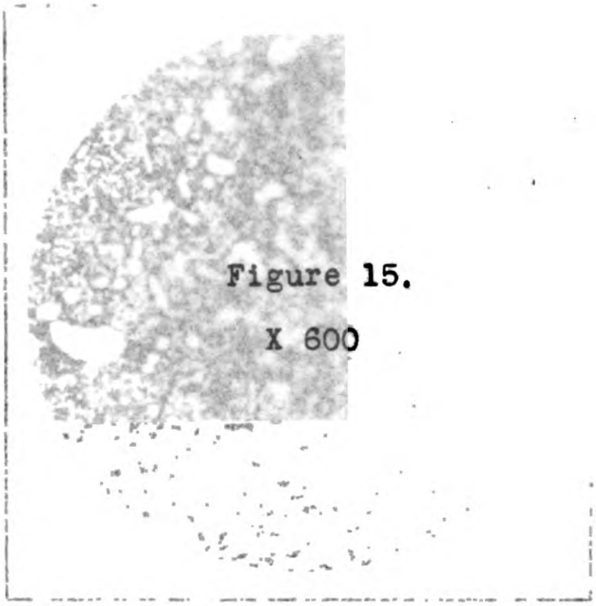
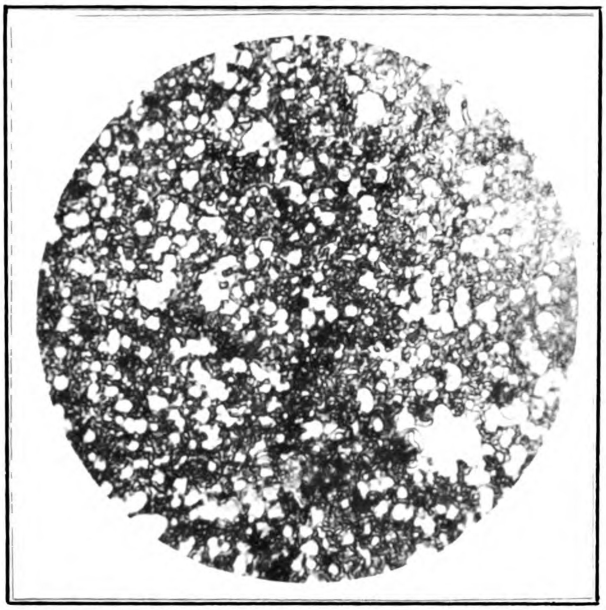
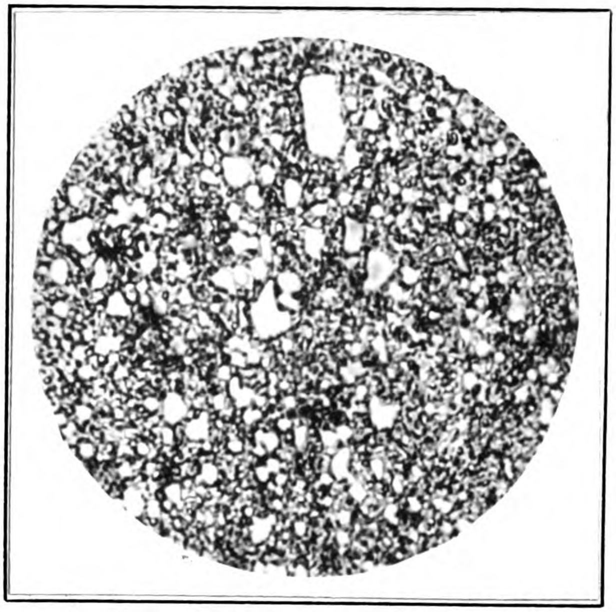


Figure 15.

X 600





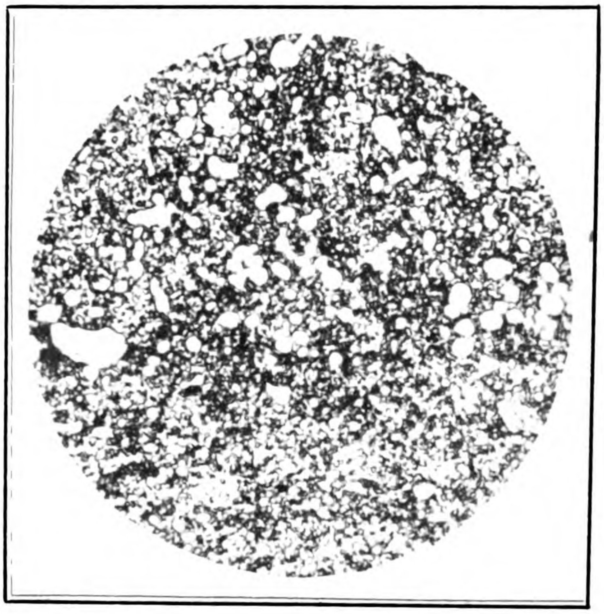
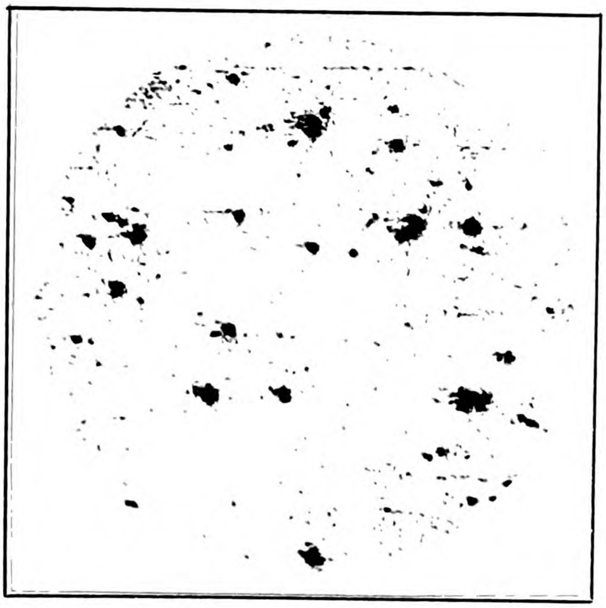


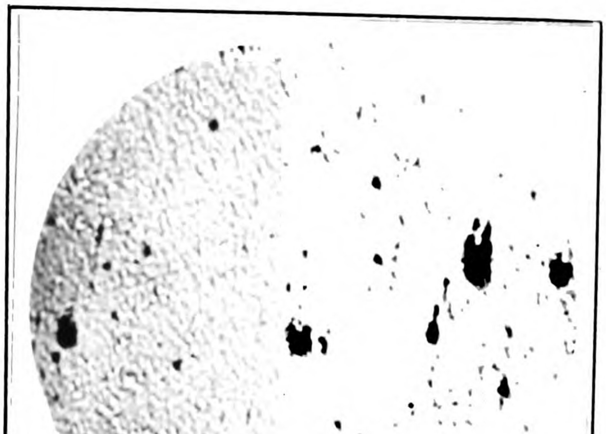
Figure 16.

X 600

Figure 17.

X 600

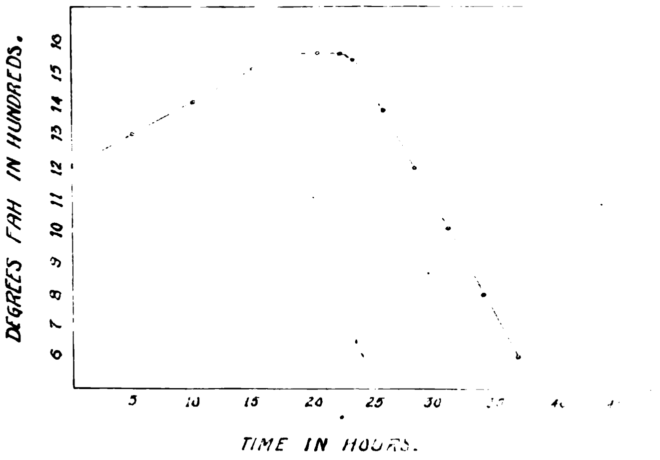




showing the sonims or slag inclusions. They are fairly small in size, well distributed and deep-seated and therefore probably not very harmful. But they do seem a trifle too plentiful for a heat of steel that finished so well in the furnace. and gave no trouble from beginning to end.

Figure 18 shows a graph of the mill anneal from which these samples were taken. It shows that the furnace was held 8 hours at temperature and then cooled 20° per hour. The anneal involved about 40 hours.

Figure 18.



ANNEALING CURVE.

SUB-CRITICAL OR WATER ANNEAL

This is resorted to only when great speed is required, such as test samples from melting furnace to laboratory during manufacture.

The method the writer has found best is: the sample on being brough from furnace is placed in a furnace recording 1500° F. and allowed to fall and equalize with steel at 1450° to 1475° F. The best results were obtained by holding sample at this temperature for 20 minutes, but where greater speed was necessary a 10 to 15 minute soak would permit drilling. After being thoroughly saturated the sample is removed from furnace and placed in a spot secluded from light rays (a pipe closed at one end serves exceptionally well). When color leaves, remove test piece from pipe and test with file until a soft point has been reached, a light yellow color shows on filed surface. At this point quench test sample in cold water, taking out quickly. Test with file and repeat as above. Cooling should be hastened after the file "softness" has been held. Total time required is less than 20 minutes.

This treatment is not a true anneal. The steel has not gone through the critical or transformation range. So in reality it is merely a very high tempering treatment.

It is advisable to allow the test sample to coal down considerably before placing in furnace.

This is probably due to the fact that above 600° to 800° F austenite remains stable, and of course if quenched from 1450° F. without first allowing a transformation to martensite at 600° F. the steel will naturally remain austenitic and undrillable.

According to Wills good results are obtained by leaving specimen in the furnace 5" at 1320° to 1400° F. This gave him a 364 B.H.N. and was drillable.

METALLOGRAPHY OF HEAT TREATMENT

The complexity of the chemical analysis of high speed steel involves an entirely different and radical departure in heat-treatment in order to obtain the desired properties, namely, maximum red-hardness.

Red-hardness as its name implies is the ability of the tool to retain a cutting edge at elevated temperatures without undergoing considerable erosion or breakage. It must be hard, yet not sufficiently brittle to cause spalling or breaking. The maximum red-hardness has been found to be directly proportional to the carbide solution, and this in turn varies directly with the temperature of quench. Such elevated temperatures as 2300° to 2400° F. are not uncommon. With such radical departures in analysis and heat-treatment it might be thought that this type of steel is different constitutionally. It is nevertheless fundamentally a hyper-eutectoid steel. A well annealed sample of this steel consists of ferrite and carbides, which compares with ferrite and cementite of hyper-eutectoid steels.

The series of photo-micrographs in figures 7 to 11 inclusive show a typical annealed structure of this steel, and consists of a matrix of ferrite (perhaps some sorbite or a modification) in which innumerable complex carbides of tungsten, chromium and iron are finely dispersed. In the final heating for

working, these excess carbides (not in solution) have been rejected or precipitated. In later photomicrographs it will be noticed that the carbides on quenching are not only in the matrix, but also right within the austenitic grains retained on quenching. This would be evidence to indicate that these particular carbides never went into solution while above the critical range.

It has previously been mentioned, see Figures 13, 14, and 15, that the annealed structure ready for heat-treatment has two distinct sets of carbide grains within the matrix, the same material but of different size and from a different source. The large granules are the remnants of the ingot eutectics existing after casting and the smaller ones are reprecipitated on cooling from working temperatures which are naturally above the critical range. This existing condition, namely, the large excess carbides is the main reason it is necessary to depart so radically from the regular heat treating methods.and quench from a temperature much beyond the transformation range.

In order to get the maximum hardness with this complex type of analysis, and , as previously mentioned obtain the red-hardness necessary for tools of this type, it has been found that maximum solution of carbides of tungsten, chromium and iron is only obtained when steel is cooled from exceedingly high temperatures.

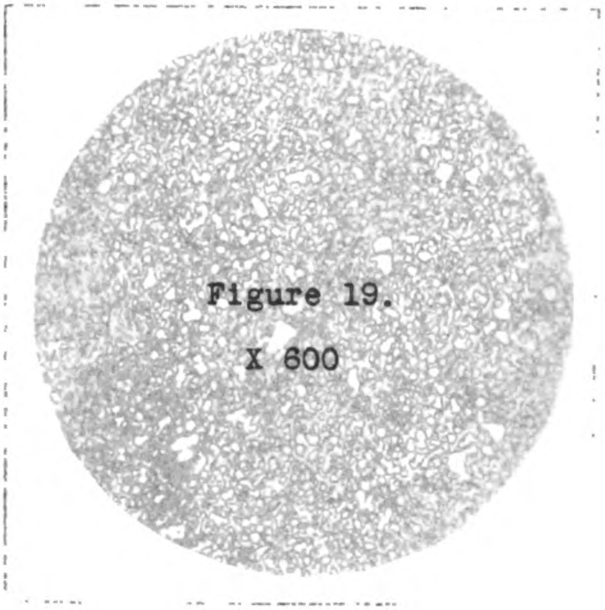
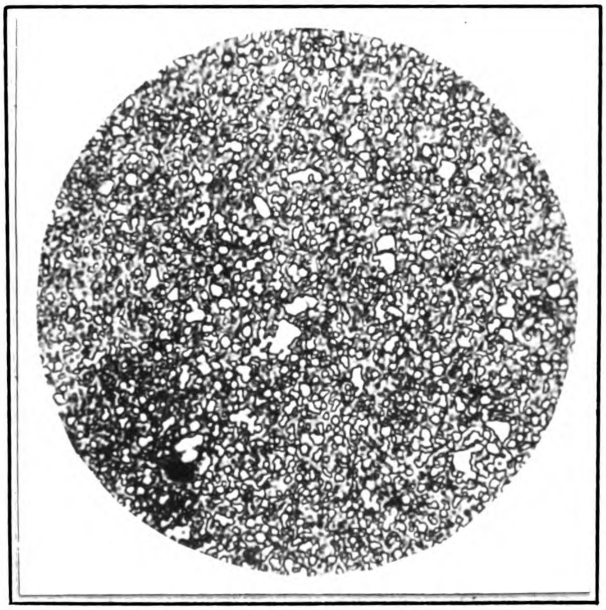


Figure 19.

X 600



It is not difficult at the temperature used for hardening, 2200° to 2400° F., to obtain solution of the secondary carbides, but the primary carbides seem to remain embedded in the hardened matrix. Heat apparently has little effect upon them.

Samples were quenched at varying temperatures starting from 1400° F, which is considerably below the critical range, and extended to a point where incipient fusion set in. An attempt was made, by the use of photomicrographs at these varying temperatures, to show the increase in solution of these precipitated carbide particles as the quenching temperature was increased. As has been mentioned, the quality or hardness of the tools increases with the increased solution of the carbide grains. There is, however, a limiting point where the brittleness and excessive grain growth resulting from overheating more than offsets any gain from the extra carbide solution. Where scale on tools is a disagreeable condition, it is offtimes necessary to sacrifice some solution (lower temperature) of carbides to retain a smooth finish to the hardened piece.

Figure 19 shows what may be regarded as the general nature of structure before hardening.

In series 20, 21, 22 and 23 are shown photomicrographs of samples hardened in oil from 1400°, 1480°, 1510°, and 1600° F.

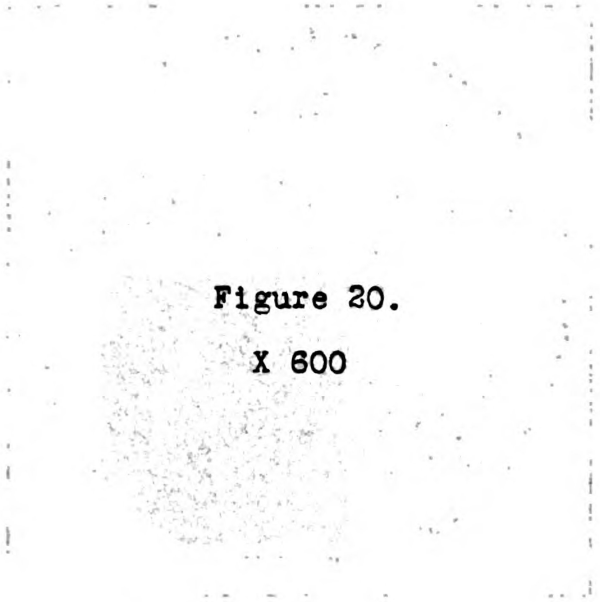


Figure 20.

X 600

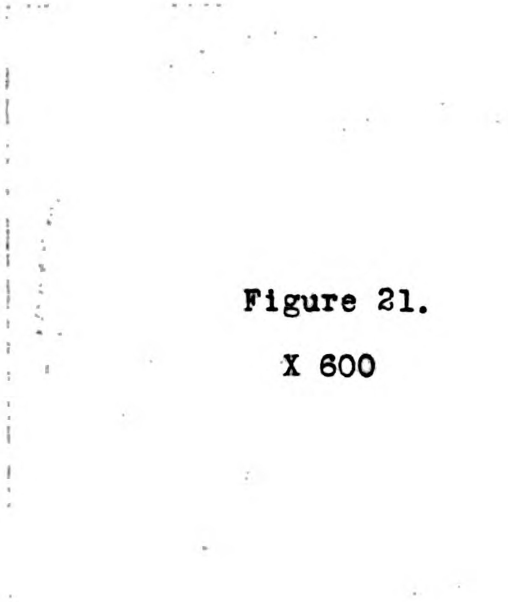


Figure 21.

X 600

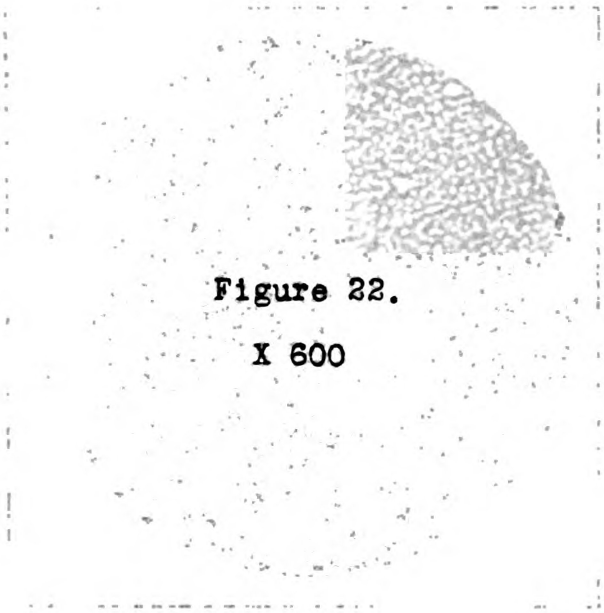


Figure 22.

X 600

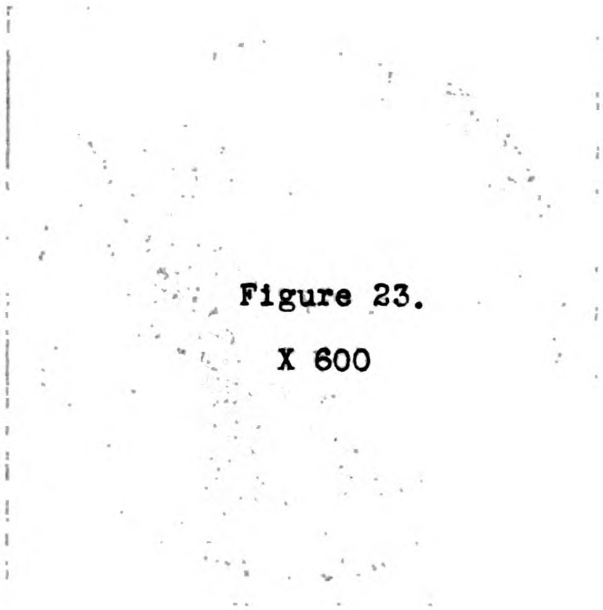
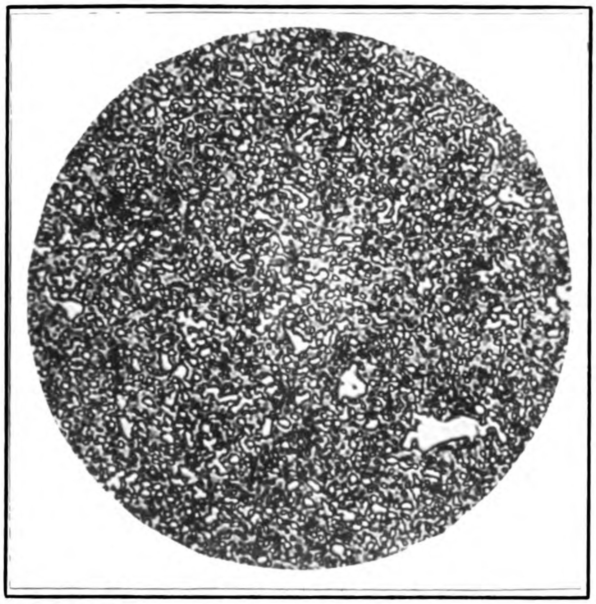
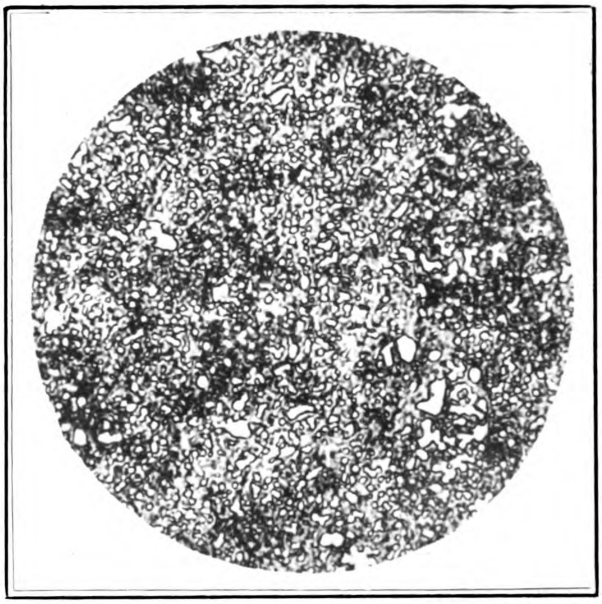
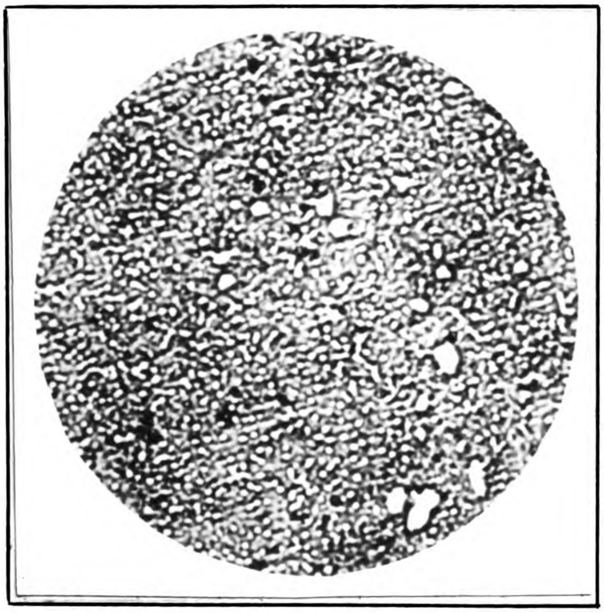


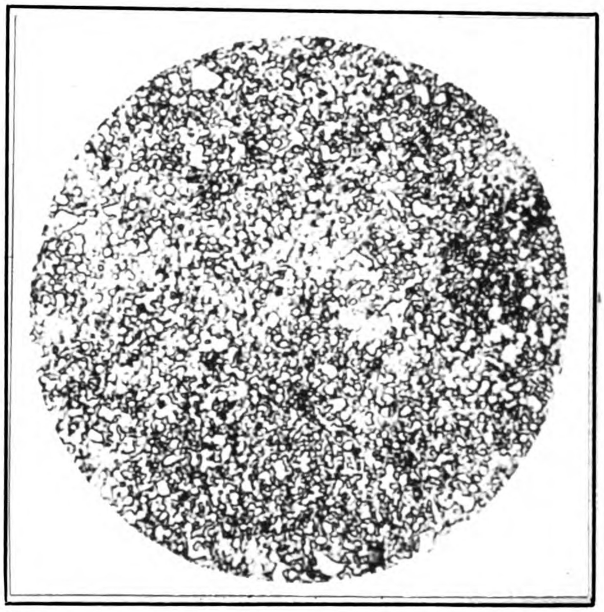
Figure 23.

X 600









respectively. All temperatures are below the critical range, yet all, with the exception of Figure 20, showed considerable solution of carbon bearing tungsten before alpha ferrite was transformed into a solid solution of gamma iron.

The Rockwell hardness of the steel before hardening was C 16. After the foregoing quenches they were C 16, C 45, C 50 and C 52 respectively. These figures show a decided hardening action. The samples used in all these testswere cut before hardening to a size suitable for microscopic examination. Thus the time required for heating and cooling was considerably shorter than is usual with tools in practice.

The sample hardened at 1600° F. required about 5" to show a uniform heating. We can compare this sample Figure 23, with one quenched from the same temperature namely 1600° F, but held for a much longer period. Figure 24 is a photomicrograph of this sample after being held nearly 15" at the above temperature. There is perhaps a little increase in carbide solution , but not nearly as great as one might expect. The Rockwell hardness was C 52 which would indicate no appreciable solution of carbides due to the lengthy soak. The photomicrograph looks as though a slightbreaking up of carbideshas started.

The series, 25 to 29 inclusive, show photomicrographs from increasingly higher temperatures with a corresponding increase shown in carbide solution and in Rockwell hardness. In Fig.26

Figure 24.

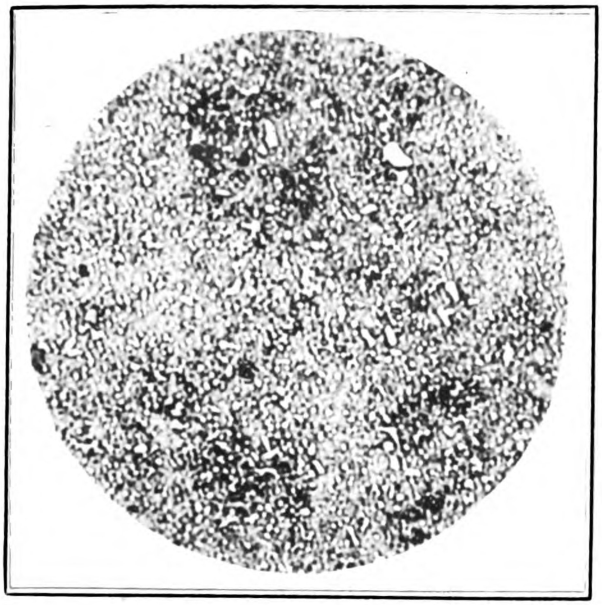
X 600

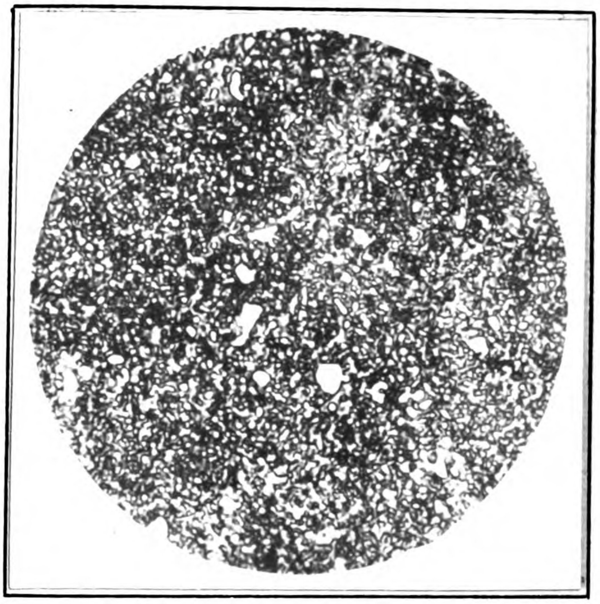
Figure 25.

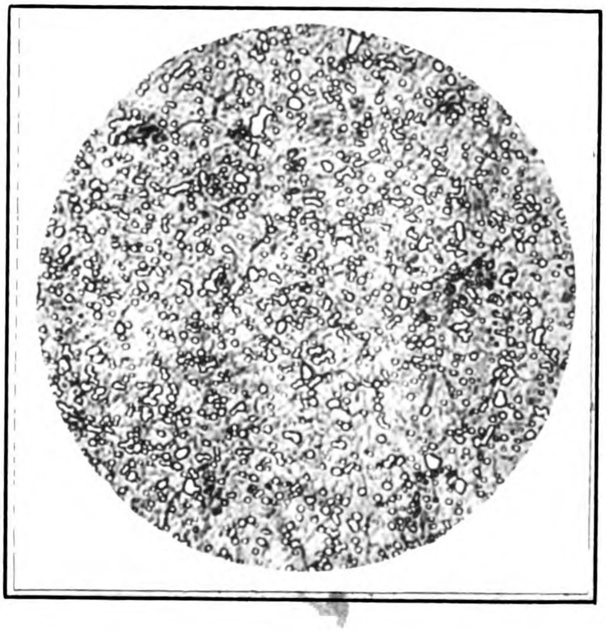
X 600

Figure 26.

X 600







we observe the first slight sign of a polyhedral (austenitic) structure.

The hardness between 2250° and 2400° F. seems to vary but little. It would seem as though maximum carbide solution, in a sample of this size, takes place at a rather low temperature. Owing to the density and sluggishness of high speed steel, it is always advisable to preheat slowly and thoroughly at a temperature around 1500° F. before bringing up to high heat. If this is not done heat-strains are set up within the tool and rupture frequently results. The tests quenched from 1800° F. and above had a preheat at 1500° F. The time held at hardening heat is shown in Figure 40.

In Figure 26, sample quenched from 2000° F, the first sign of an austenitic structure is evident. The samples quenched from the higher temperatures show this characteristic to a more marked degree with a much larger growth of the individual grains.

In Figures 30, 31, and 32 the photomicrographs show the specimen at increasing magnifications, after being hardened at 2400° F. Figure 32 brings out clearly the austenitic condition.

Figure 34 is the same sample as Figure 33 with the exception of being reground considerably the second time on account of the length of time sample was held at the high temperature.

Figure 27.

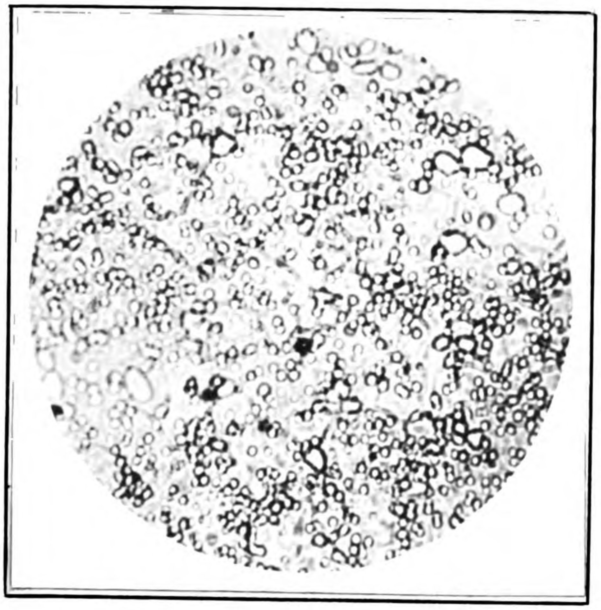
X 600

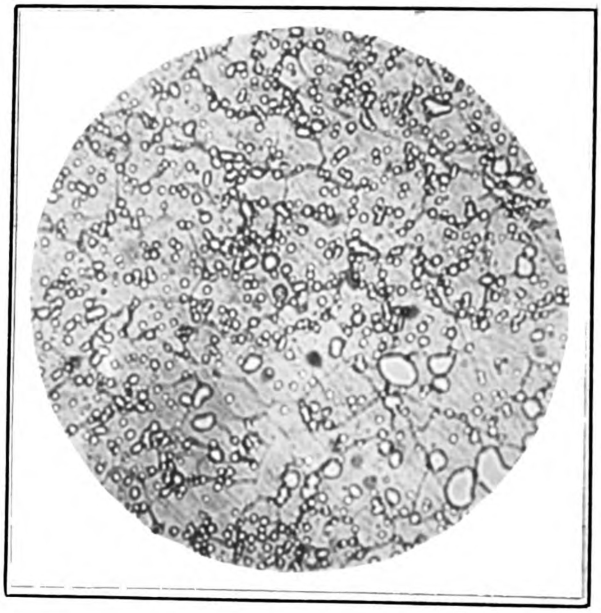
Figure 28.

X 600

Figure 29.

X 600





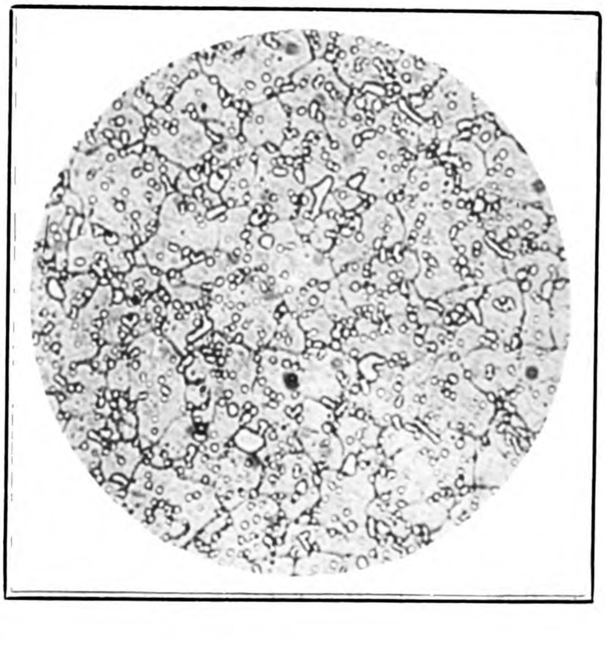


Figure 30.

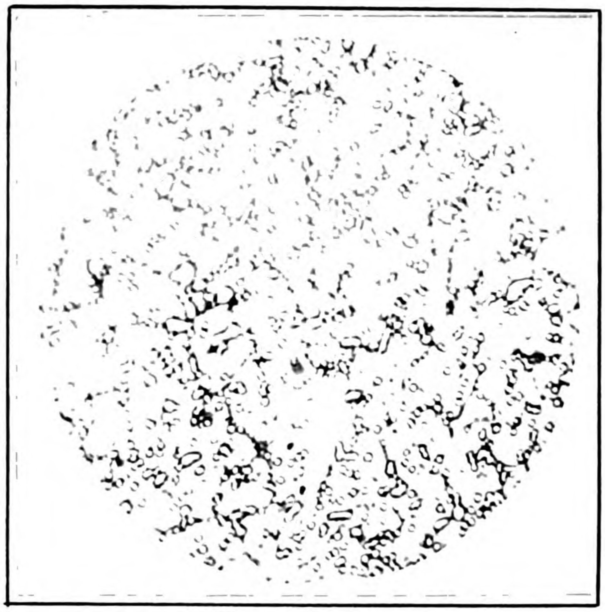
X 600

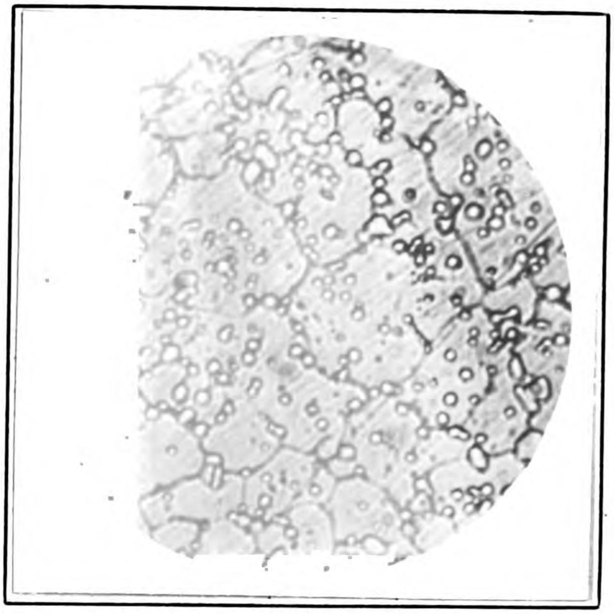
Figure 31.

X 1200

Figure 32.

X 2400





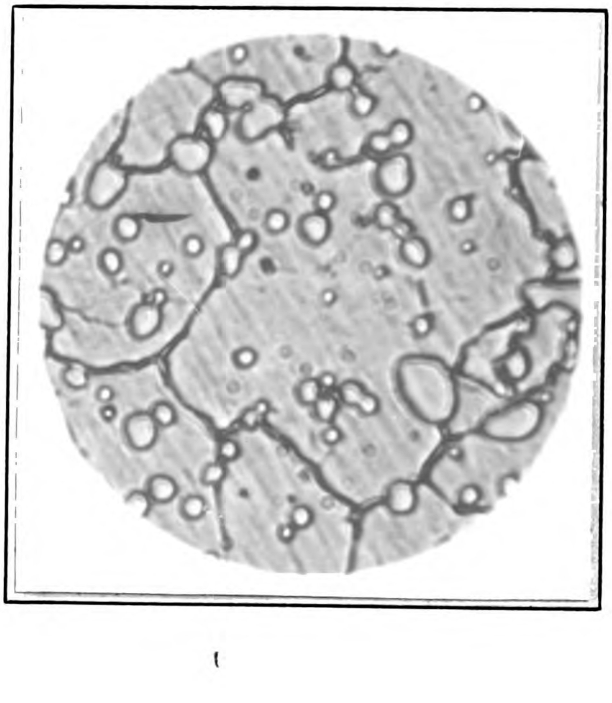


Figure 33.

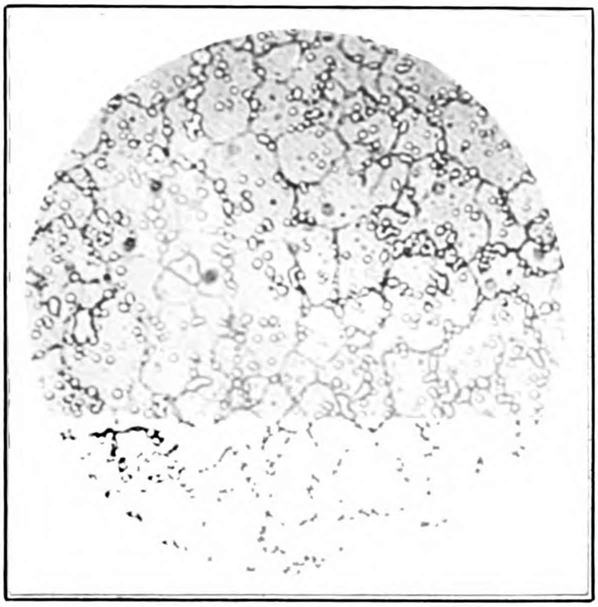
X 600

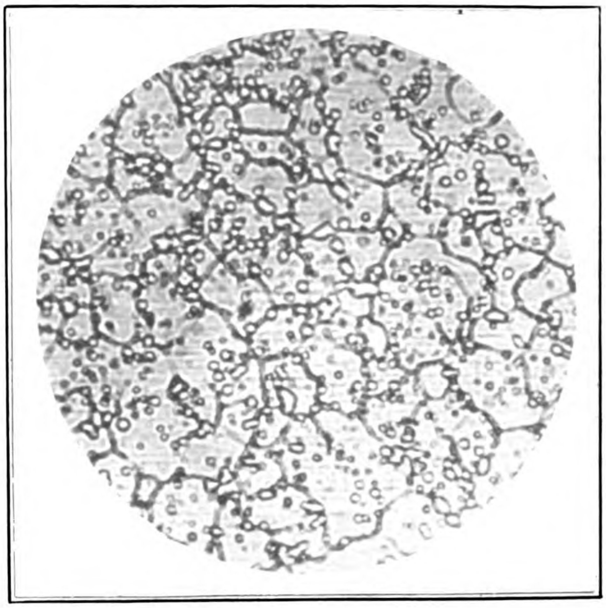
Figure 34.

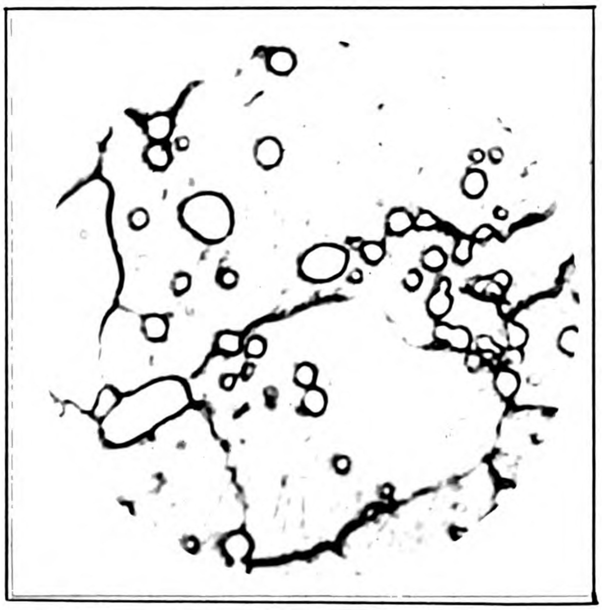
X 600

Figure 35.

X 2400







The only noticeable change is an increase in the width of the grain boundaries. Figure 35 is the same as Fig. 34 with increased magnification.

Sample in Fig.35 was quenched from the same temperature as Fig.32 only instead of quenching after 1¼ minutes, when it showed complete saturation of heat,it was held 4 minutes to observe the effect on the solution of carbides. The effect,except for enlarged grain size,is negligible. In all probability such a tool would show an increase in brittleness. The Rockwell hardness remained approximately the same.

Photomicrograph shown in Fig.38 is a sample heated in a furnace recording a temperature of 2475° F. It was held 2 minutes and on removal showed excessive scale in the form of large "sweat" blisters. The sample was decidedly overheated and held too long. The Rockwell hardness was C64 which is much below those samples hardened in the regular manner. The grain growth in this specimen is very large,and it may be possible that excessive decarburization at this elevated temperature has brought about the precipitation of some ferrite. Figure 39 is a similar sample except that it was run at a temperature (2500° F.) where incipient fusion has set in. The characteristic eutectic structure ordinarily found in the solidified steel after ingot cooling,has now been produced by overheating. It is

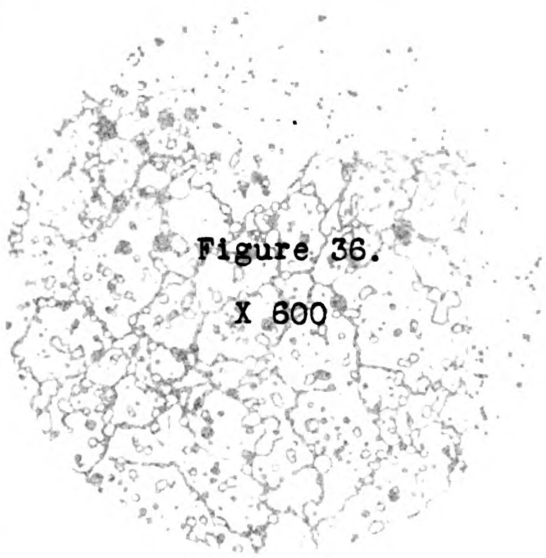


Figure 36.

X 600

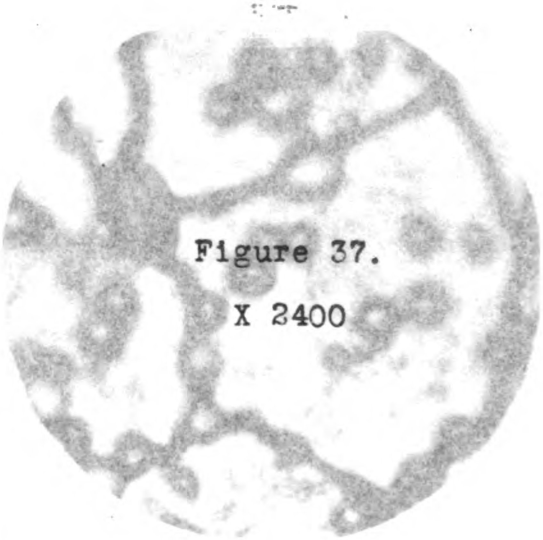
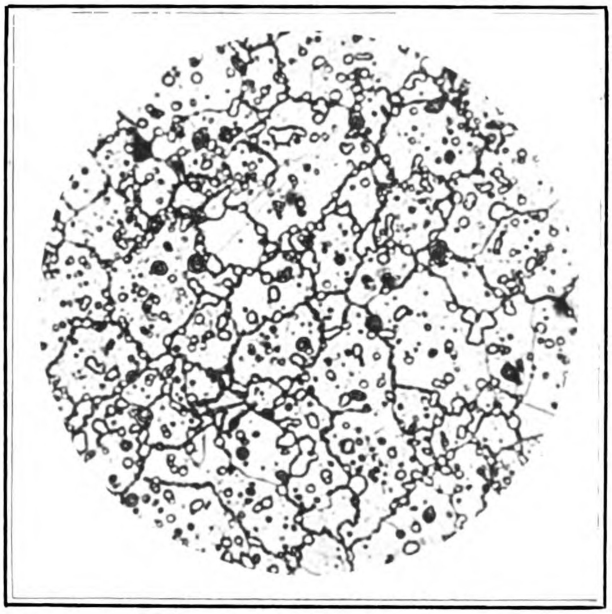
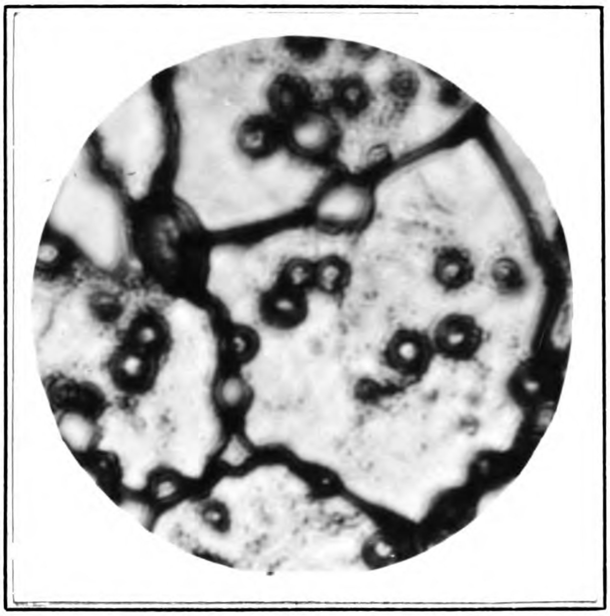


Figure 37.

X 2400





quite probable that such a temperature causes decarburization and ferrite precipation. Austenite is not stable in solid solution if the minimum carbon percentage is not maintained. This is a typical structure of badly overheated steel. The structure that etched dark in Fig.38 as well as Fig. 39 is in all probability precipatated ferrite resulting from carbon diffusion. Such a condition as shown in these two samples is very undesirable; it induces brittleness and weakness. The samples shown in Figures 36 and 37 show the polyhedral structure characteristic of austenite which has been retained in the solid solution after the specimen was cooled in still air from 2350° F. This indicates that the speed of cooling is not a factor in the retention of carbides in solution. The Rockwell hardness is C. 66 and compares very favorably with those quenched in a more drastic medium. In Figure 37, which is the same as Figure 36 only at increased magnification, it looks as though within the grains there is a dark constituent that might possibly be martensitic in nature. It is exceedingly difficult to etch this in a manner to show the characteristic needle-like structure of martensite.

In Figure 40 is the detailed treatment given to the various samples with their corresponding Rockwell hardness figures.

Figure 38.

X 2400

Figure 39.

X 2400

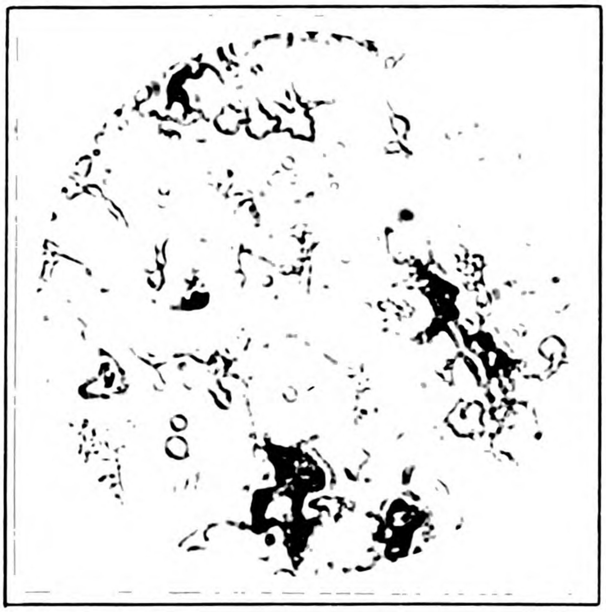


Figure 40.

Number	Pre-heat	High heat	Time at Temp.	Rockwell C
20	-	1400°F	10'5"	18
21	-	1480	10'0"	44/45
22	-	1510	8'0"	49/50
223	-	1600	5'0"	50/51
24	-	1600	15'0"	50/52
25	1500°	1800	4'15"	57/58
26	"	2000	2'5"	63½
27	"	2260	2'20"	66/66½
28	"	2310	2'0"	66½/67
29	"	2350	1'40"	67
30-32	"	2400	1'25"	67
33-35	"	2400	4'0"	66/67
38	"	2475	2'0"	63/64
39	"	2500 +	1'18"	-
36-37	"	2350 (air)	1'55"	65/66

In Figure 41 is shown the plotted hardness values at increasing temperatures. It is self-explanatory.

In Figure 42 we have the corresponding values after increasing tempering temperatures. It will be noticed that with those samples fully hardened the maximum secondary hardness is obtained between 1000° and 1100° F. This temperature also gives the maximum red-hardness.

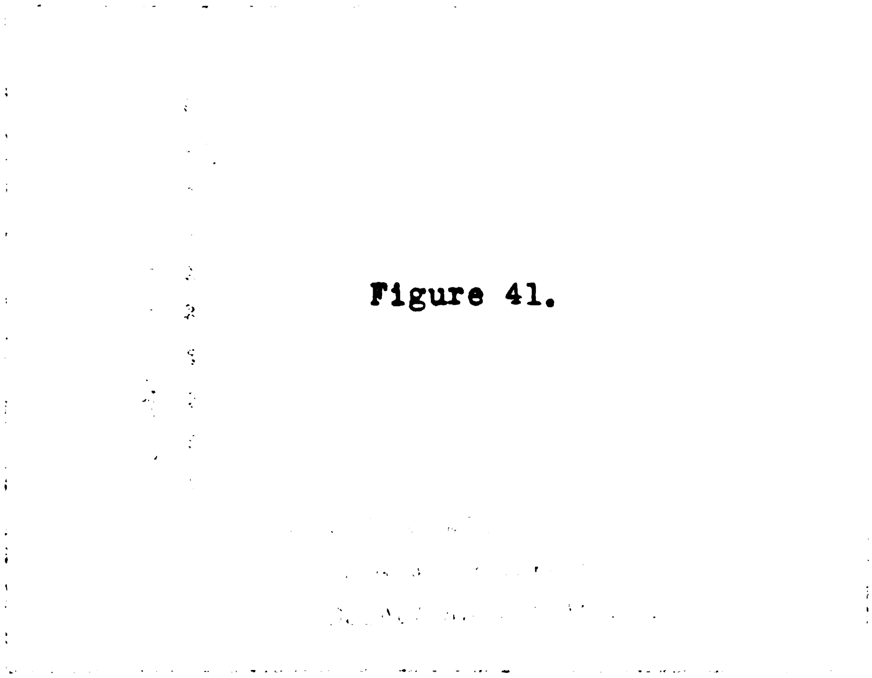
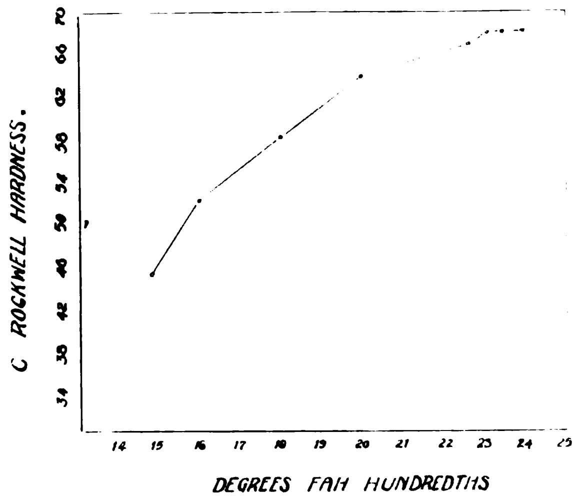
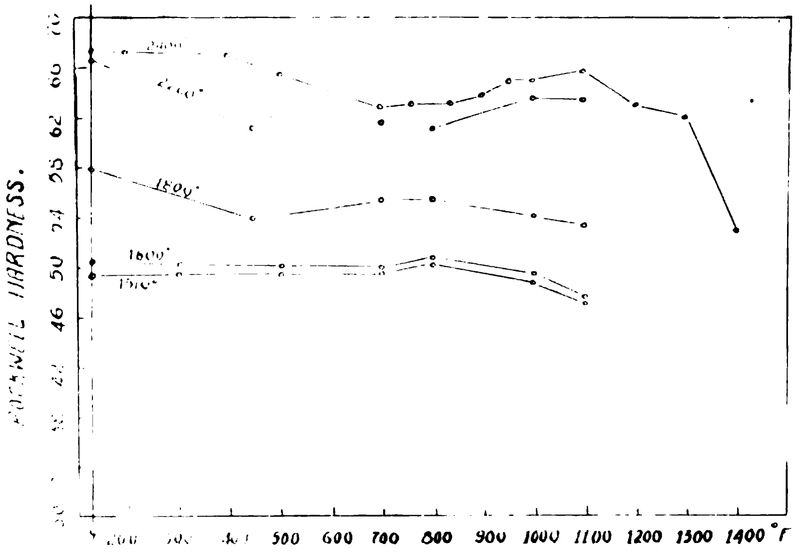


Figure 41.



QUENCHING HARDNESS.

Figure 42.



TEMPERING CHART.

TEMPERING OR DRAWING

Paradoxical as it may sound the drawing or tempering of high speed steel at the usual temperature, namely, 1000° to 1100° F. results not in a reduced hardness, as is usual with carbon steel, but in a decided increase in hardness. This is generally referred to as secondary hardness, and represents the maximum red-hardness. This condition is shown graphically in Figure 42.

The sample used for all tempering operations was that shown in Figure 32 and was quenched in oil from a temperature of 2400° F. after a 1500° F. preheat.

In the quenched condition it is seen to be typically austenitic in structure. The polyhedral grains are well defined with the primary carbide grains mixed within the grains and along the grain boundaries. This excess carbide dispersion and distribution may be quite a factor in the hardness and heat stability of this type of steel. Its nature would diminish the atomic mobility and increase sluggishness.

The first few specimens tempered are clear and easily examined at low magnification, hence the draws between 700° and 900° F. were photomicrographed at 600 diameters. Above this temperature 2500 diameter magnification permitted a better interpretation of the changes taken place.

Figure 43.

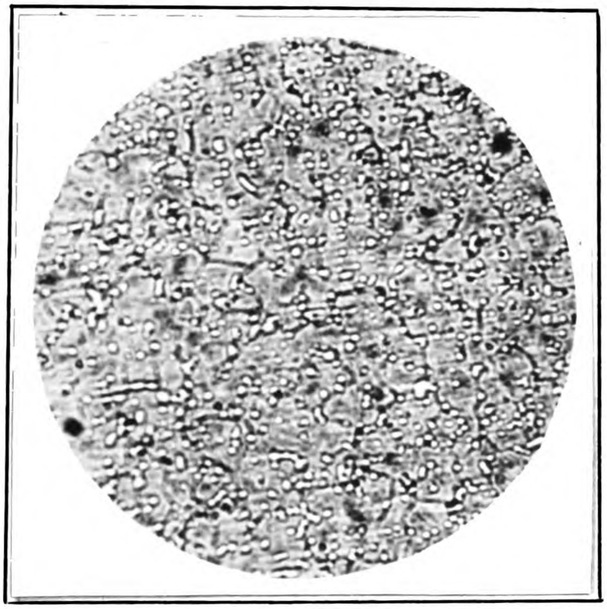
X 600

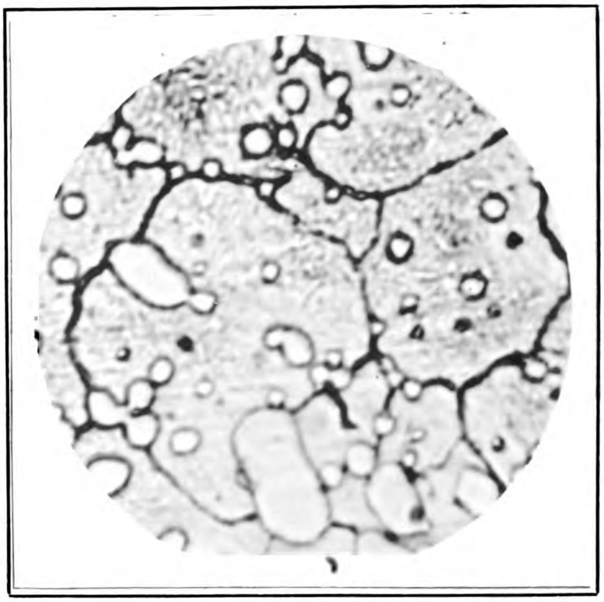
Figure 44.

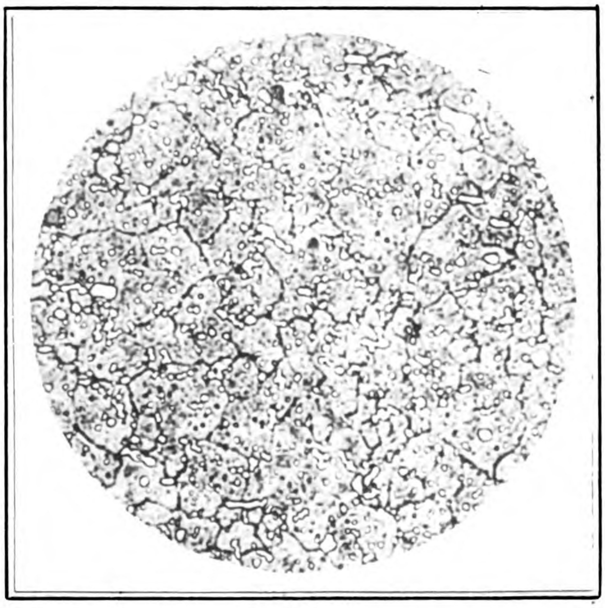
X 2400

Figure 45.

X 600







The tempering operation was started at 700°F, a point well below the martensitic transition point. No apparent change was noticed when etched with the regular 5% Nital reagent. The Rockwell hardness dropped sharply from C 67 to C 62½. This would indicate a physical change, probably the removal of hardening strains from within the specimen.

All samples were drawn for a period of one hour at temperature. This should be sufficient time to complete any transformation in a sample of this size.

Figure 43 shows the photomicrograph of sample after a draw at 760°F. for one hour. This is apparently right at the point where a start is made in the transition of residual austenite to the more stable and harder constituent martensite. This is well born out by the Rockwell hardness increasing from C 62½ to C 63½. No change is apparent in the photomicrograph, but this is probably due to the difficulty involved in showing martensite or martensitic structures with ordinary etching reagents.

Figure 44 shows the same sample after tempering at 840°F. It seems strange that no increase in hardness is shown with this extra tempering temperature. The Rockwell hardness remains C 63½. The photomicrograph shows no great change, but with heavy etching a slightly darkened condition is noticed within the austenitic grain boundaries. An increase in magnification was used to make resolution more

Figure 46.

X 600

Figure 47.

X 2400

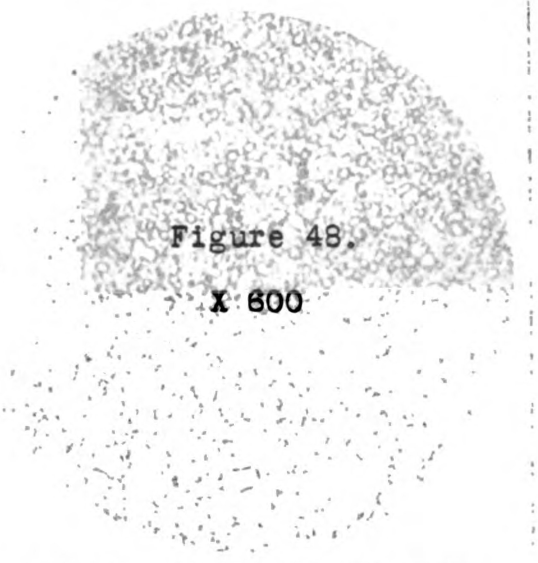


Figure 48.

X 600

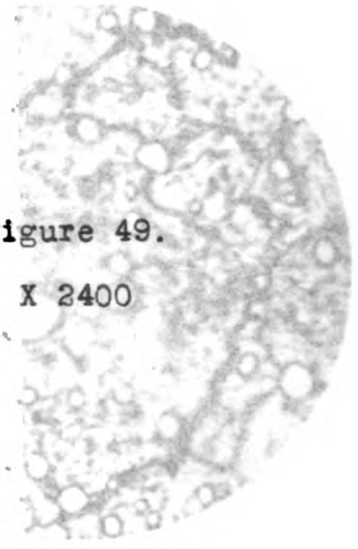
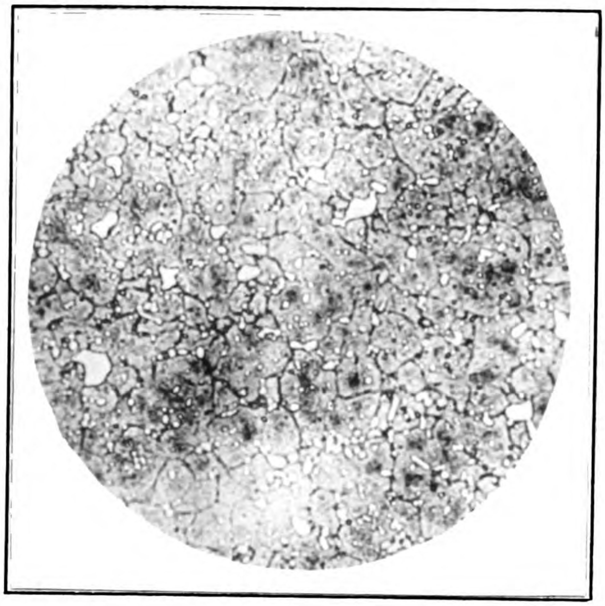
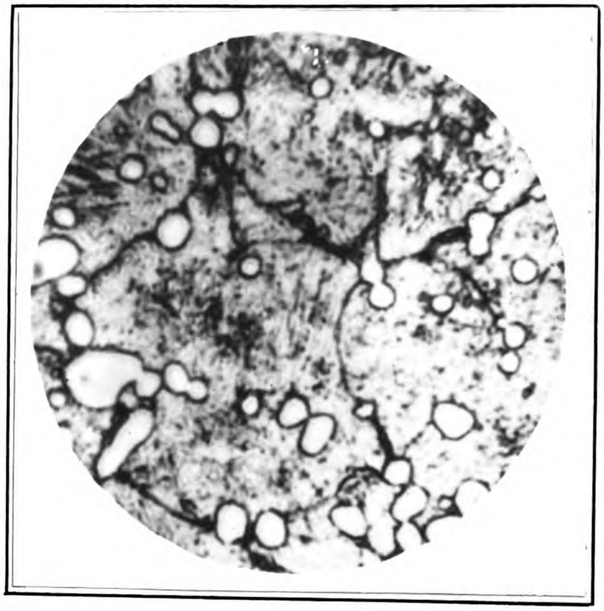
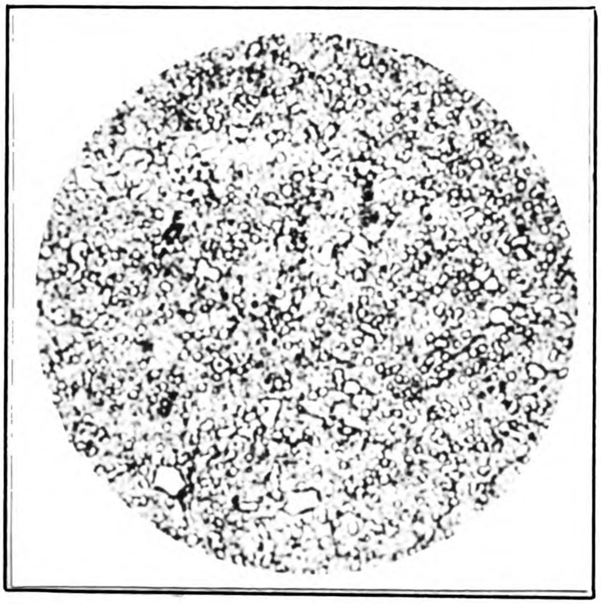


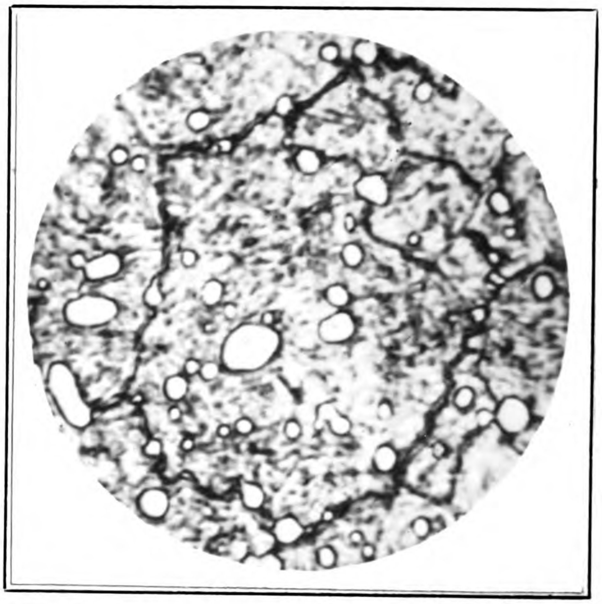
Figure 49.

X 2400









certain. This may or may not be a martensitic condition; it is very difficult to say with any degree of accuracy.

In Figure 45 the sample has been drawn at 900°F. The Rockwell hardness is now C 64, a slight increase. The grain boundaries do not appear to have changed, they are clear and well defined. The sample showed a slight tendency to etch faster.

Figures 46 and 47 show the sample after a 950°F. draw. The Rockwell hardness showed a small increase to C 65. The Matrix within the grain boundaries shows a decided transformation or decomposition. The polyhedral grain boundaries stubbornly persist, and show very little, if any, effect from this temperature.

In Figures 48 and 49 the sample has been drawn at 1000°F. The Rockwell hardness remains C 65 which is very close to the original hardness on quenching. The original polyhedral grain boundaries are not as well defined as in former pictures, but they can be traced with difficulty. The etch in this case was slightly too deep as it expressed the nature of the matrix, which somewhat obliterates the definition of the grain boundaries proper.

Figures 50 and 51 show the sample after a draw at 1100°F. The Rockwell hardness is now C 66. The sample is considerably over-etched, it being difficult to find a suitable

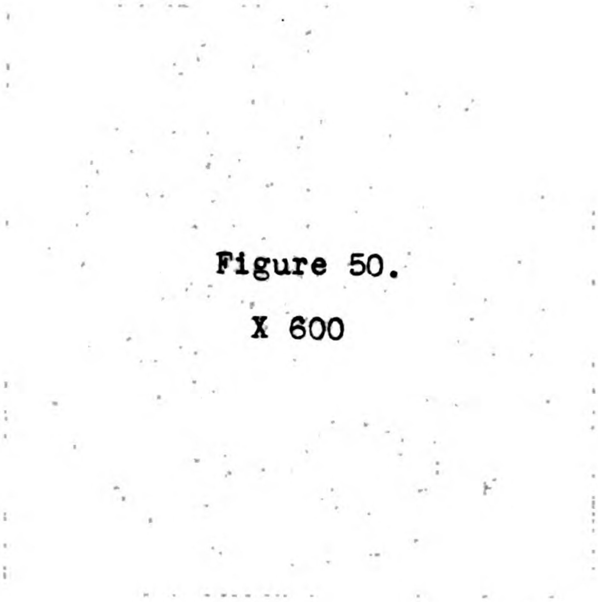


Figure 50.

X 600

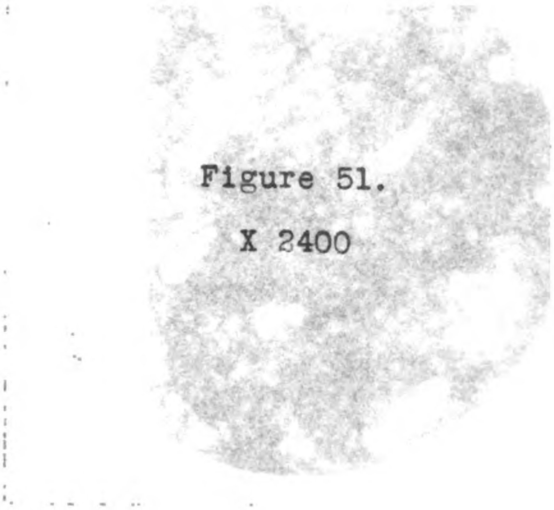


Figure 51.

X 2400

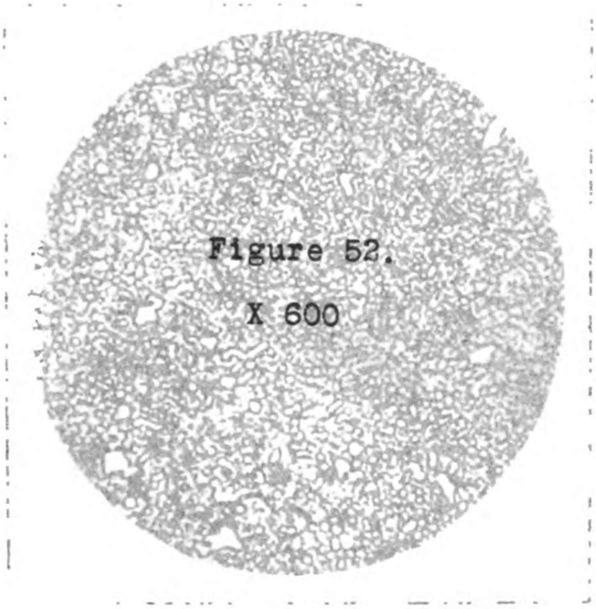


Figure 52.

X 600

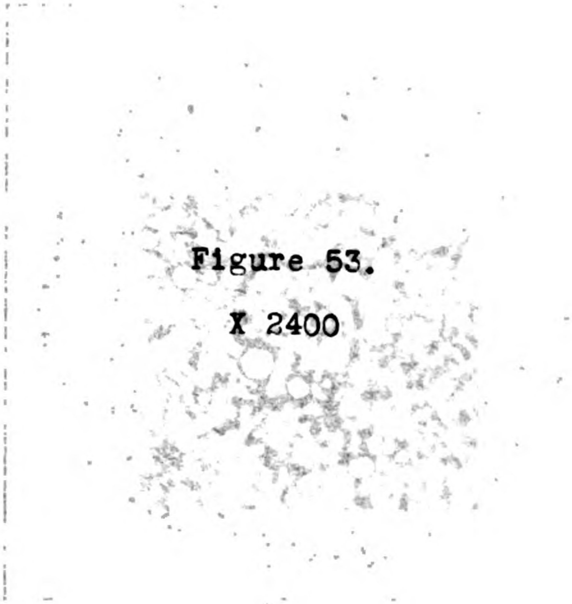
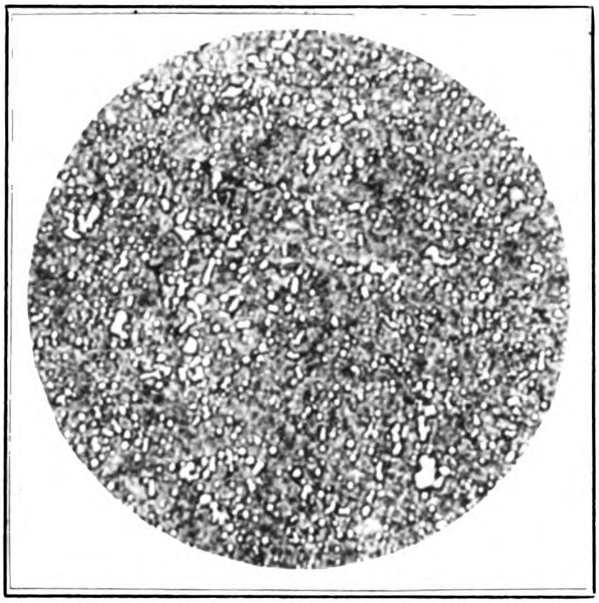
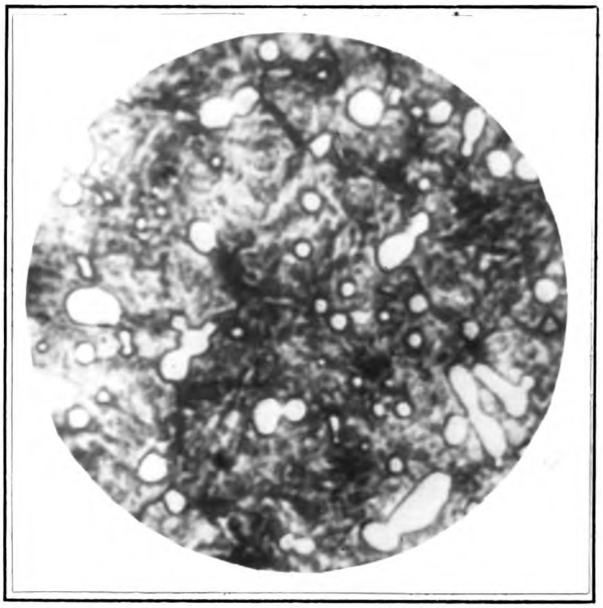
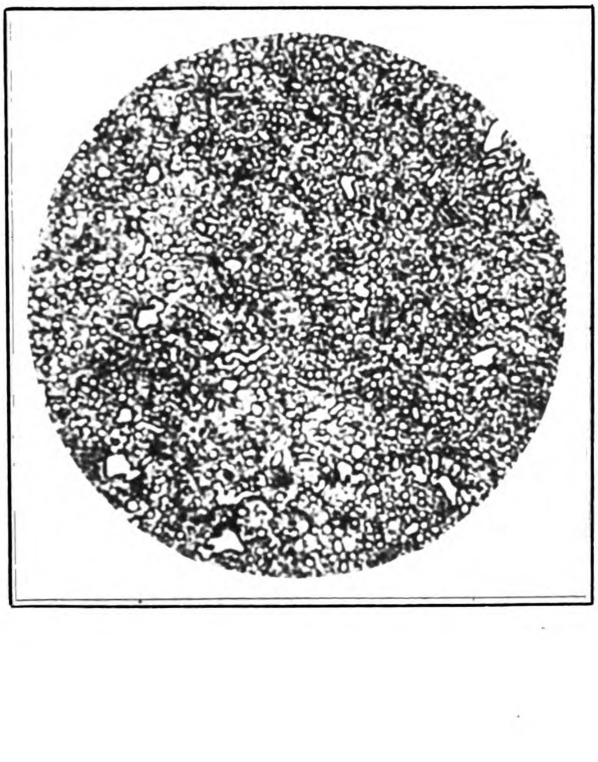


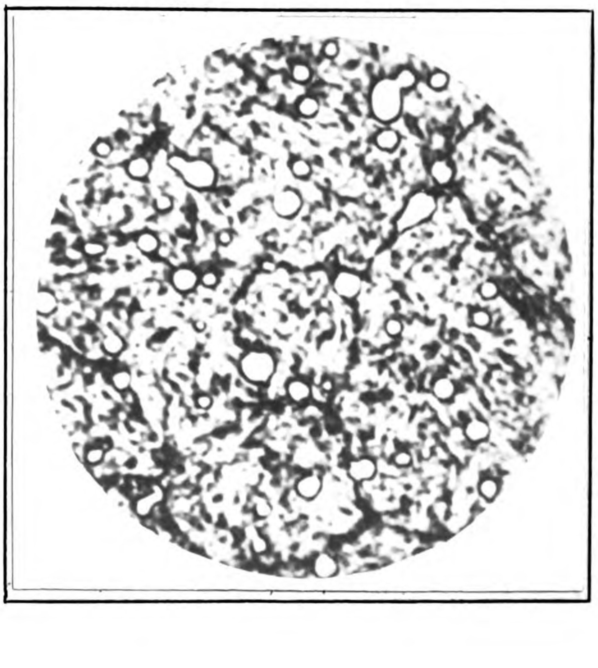
Figure 53.

X 2400









"etch-time" with Nital to resolve the grain boundaries before the matrix darkens. However, the grain boundaries are still faintly defined. It should be mentioned that the sample was given two separate draws at this temperature to make certain ample time was given at the tempering temperature. No microscopic difference was observed with the second draw.

Figures 52 and 53 show the sample after a 1200°F. draw. This is much clearer on focus and etch than the preceeding Figure, and shows clearly the well defined polyhedral grain boundaries. The Rockwell hardness of C 62-64 shows a decided drop which would indicate a second phase transformation has commended. It would probably be some form of troostite or sorbite stable at this temperature. It points out clearly the tenacity and permanency of crystal grain boundaries. The etching solution used for this particular photomicrograph was less than one percent Nital.

Figures 54 and 55 show the sample after a 1300°F. draw. The Rockwell hardness is C 61/63. Under visual microscopic examination the grain boundaries are still in evidence though not so clearly as the preceeding samples. Figure 55 requires close examination to detect the grain boundaries. The Matrix etches dark so quickly that it has become increasingly difficult to etch in such a manner as to define the grain boundaries.

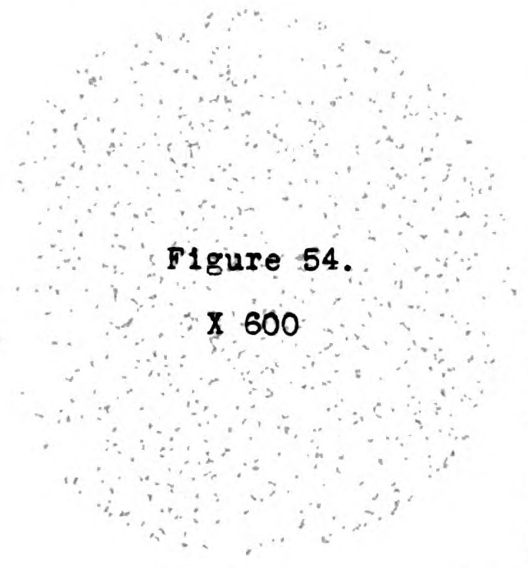


Figure 54.

X 600

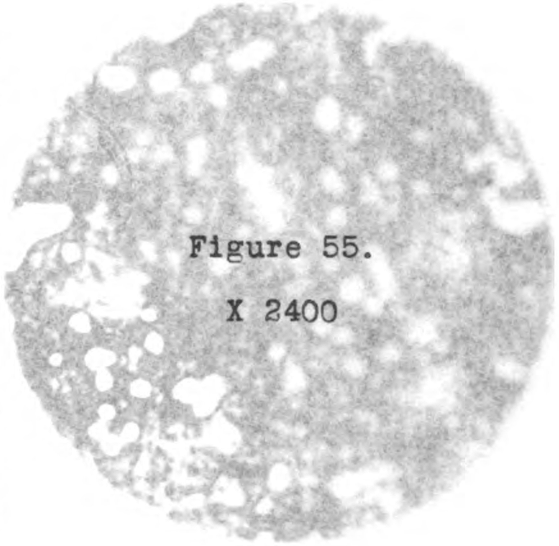


Figure 55.

X 2400

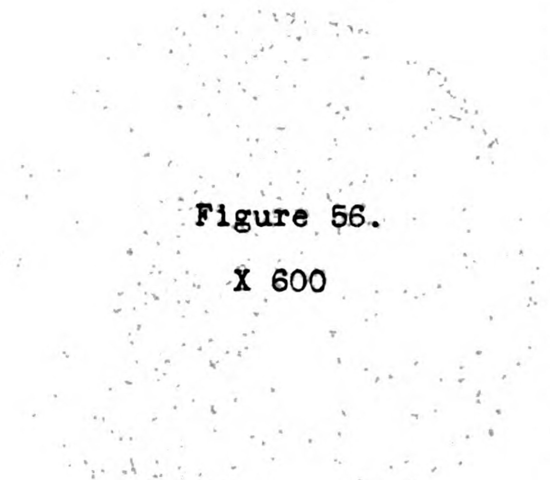


Figure 56.

X 600

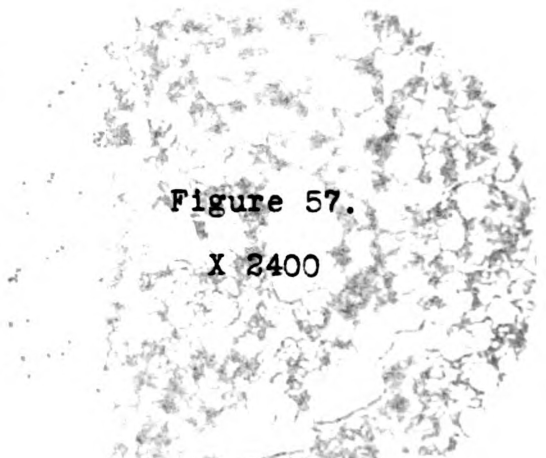
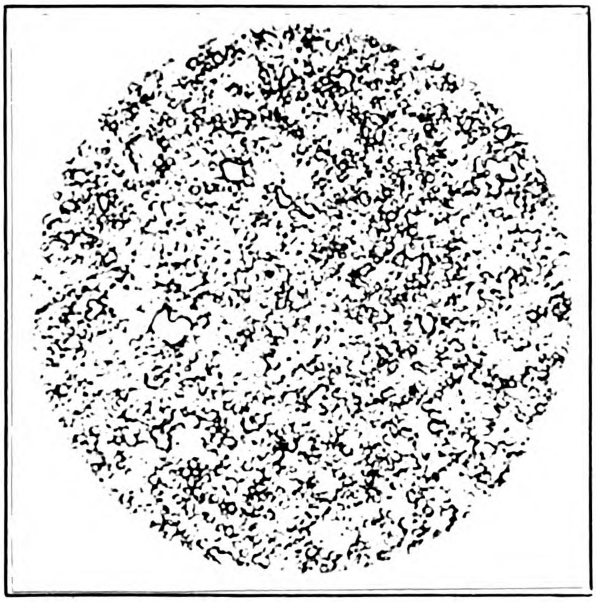
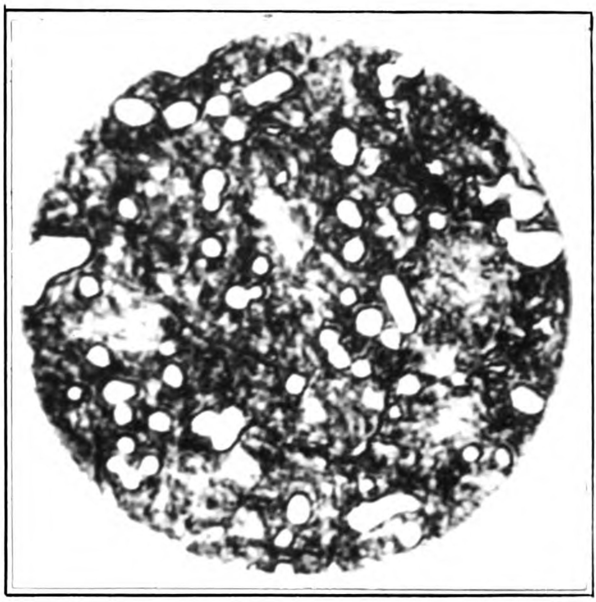
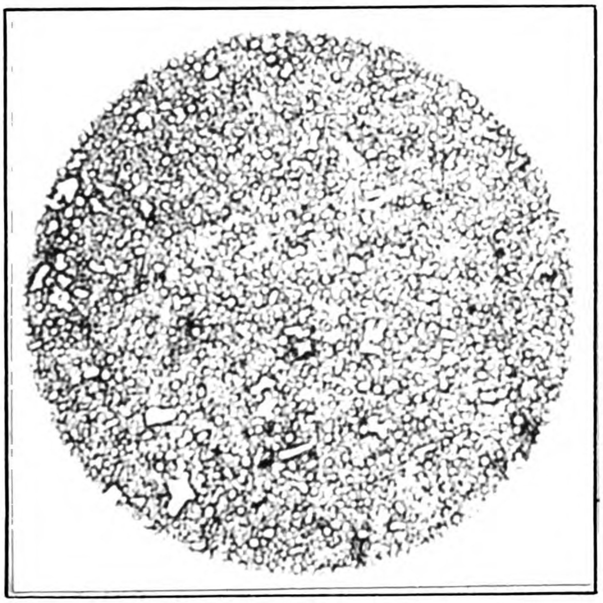


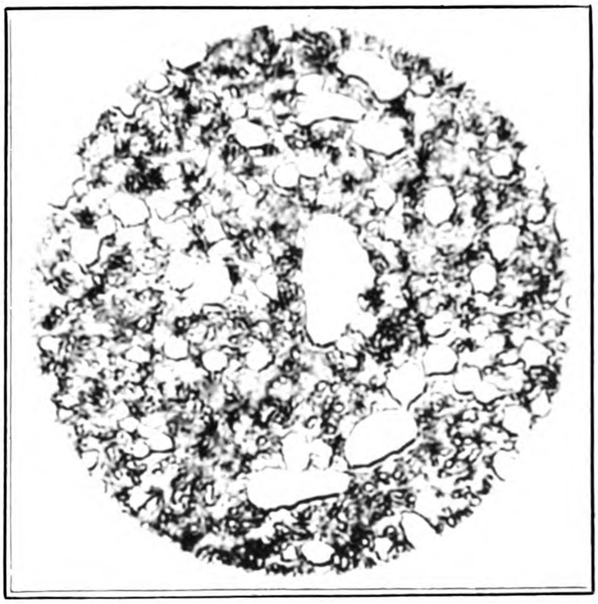
Figure 57.

X 2400









Figures 56 and 57 show the nature of the steel after a 1400° F. draw. The Rockwell hardness has dropped to C 52/53. The grain boundaries are no longer discernable, and the matrix appears to have changed considerably. There is also some evidence of carbides sperodizing.

By reference to Figure 42 will be seen the graph hardness figures of several specimens quenched and drawn from varying temperatures. This is self-explanatory.

COMMERCIAL HARDENING

The generally accepted practice for hardening tools made from the regular 18:4:1 type of high speed steel varies but little from the original treatment as recommended by Taylor and White.

The treatment should consist of four stages, namely, preheating, high heating, quenching and tempering. All stages with the possible exception of the quenching are of the utmost importance to the life of the tool.

Preheating: Owing to the extremely high temperature required for hardening and the extreme density and sluggishness of this type of analysis, it is of the utmost importance that a slow and uniform heating be given tools made from this steel. This slow heating should be adhered to rigidly until the steel has at least passed through its critical range, 1480° to 1490° F. Distortion and rupture are usually the result of excessive strains set up by heating too quickly through the transformation range. Owing to the sluggishness of the austenitic transformation in this steel, it is advisable to hold quite some time at the preheating temperature to insure complete transformation. A slightly low temperature in preheating prevents scaling of tools. A thorough preheat then permits rapid heating on transferring to the high heat furnace. A slow heating and a fairly generous soak during the preheating gives some solution of carbides as shown by the Rockwell hardness values in Figure 40.

High-heat: Quenching in general depends on the size and nature of the tools to be hardened. Small and intricate pieces necessarily demand extreme care and somewhat lower temperatures than larger sections. .In general, a temperature range of 2250° to 2400° F. is usual. The high heat furnace should be at the desired temperature before transferring the tools from the preheat furnace. This ensures fast heating and less scale. When the tool reaches the temperature of the furnace no great time is lost before quenching, as if held at this point for any length of time excessive scaling and grain growth with its subsequent brittleness takes place. With large sections there is probably always a differential in temperature of the core and the outside surface of tool. This condition cannot be eliminated as the working surface of the tool must be protected and given first consideration. The transformation to martensite on cooling takes place rather slowly, so the differential in temperature referred to above is not of very great importance.

Quenching: The speed of quenching and the nature of the quenching medium is of no great importance. It is the usual practice to quench in a thin soluble oil, but many tools are cooled in air blast or salt bath. Taylor and

White quenched in molten lead at 1100°F. After tool reaches temperature of lead it is cooled quickly to or near atmospheric temperature to effect the austenitic-martensitic transformation. There seems to be little or no gain in any of the special quenches over that of air or oil.

Tempering: This stage, along with careful heating, is probably the most important phase of well treated high speed steel tools. The usual draw is 1000° to 1100° F, where the maximum secondary and red-hardness is obtained. The length of the draw varies with different shop practices and different tools from one hour to twenty-four hours or longer. It is safe to say that no tool made from steel of this type has ever failed from too long a draw. When heavy and intricate sections are to be tempered, it is advisable to warm up slowly to the tempering temperature to prevent rupture from heat strains. The tool after quenching and before drawing is naturally in a very strained state.

By referring to Figures 36 and 51 we will see a typical hardened undrawn and drawn speciment respectively.

COMPARATIVE ETCHING.

In order to compare the structure with different etching reagents two samples,one annealed and one hardened,were selected for examination. The annealed sample was taken from bar stock which was annealed as shown in Figure 18. The hardened sample was quenched in oil from 2350° F and showed a Rockwell hardness of C 67.

All photomicrographs were taken at 2500 diameters.

Figures 58 and 59 are the annealed and hardened specimens respectively after etching with an alcoholic solution of picric acid. The annealed sample was etched six minutes, while the hardened sample required ten minutes.

Figures 60 and 61 are the annealed and hardened specimens respectively after etching with a solution of hydrogen peroxide and sodium hydroxide. The annealed sample was etched twelve minutes,while the hardened sample required eighteen minutes.

Figures 62 and 63 are the same as the two previous photomicrographs (60 and 61) after repolishing and re-etching. The tungstides are shown as much larger globules and greater dispersion through out the entire mass.

Figures 64 and 65 are the annealed and hardened samples respectively after etching with potassium ferricyanide and potassium hydroxide. The annealed sample was etched five

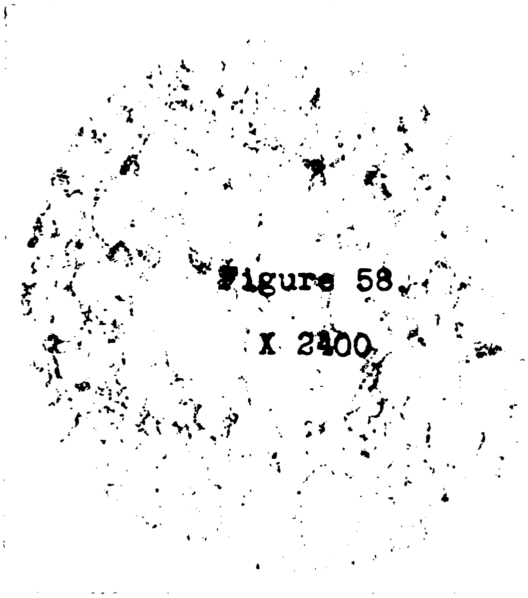


Figure 58.

X 2400

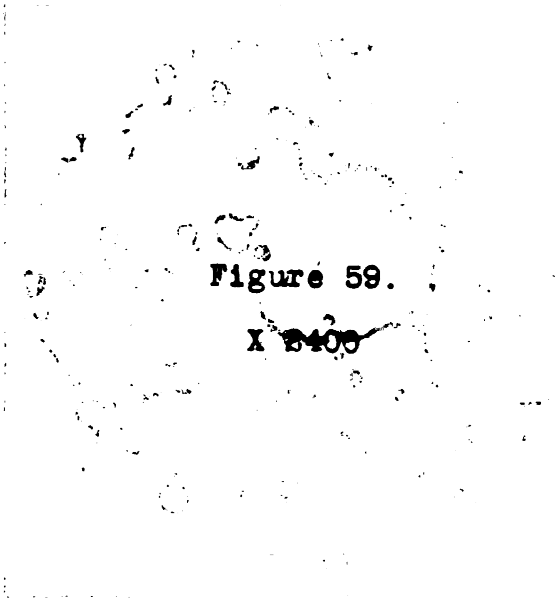
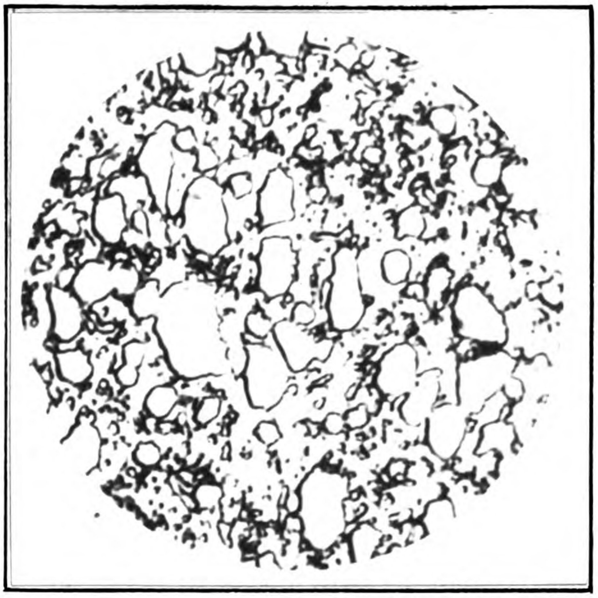
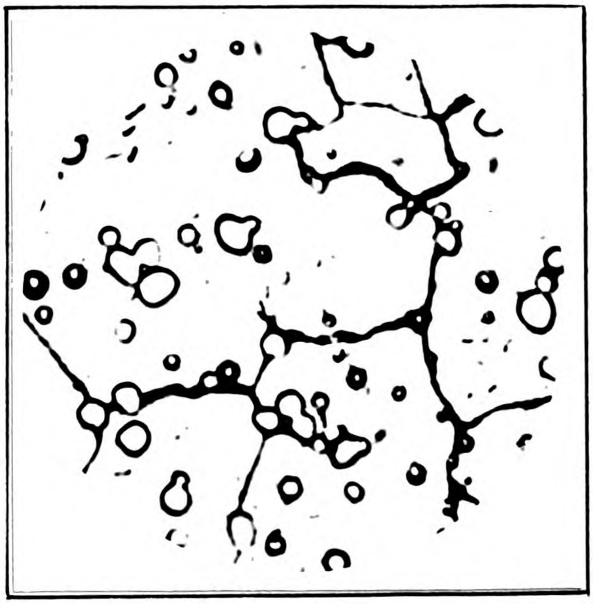


Figure 59.

X 2400





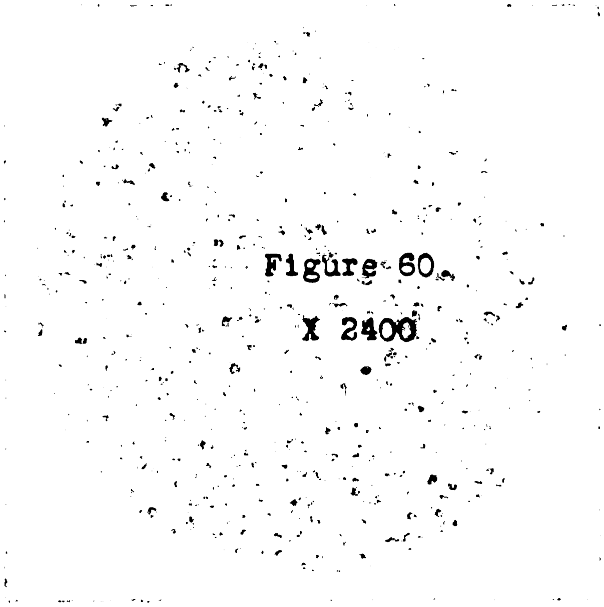


Figure 60.

X 2400

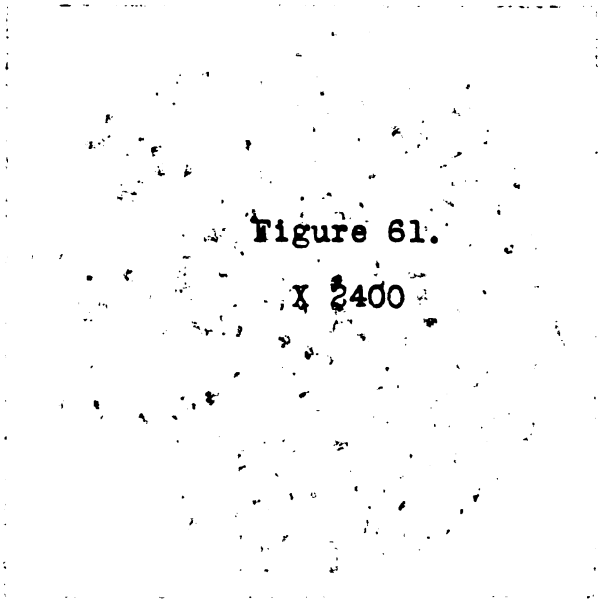


Figure 61.

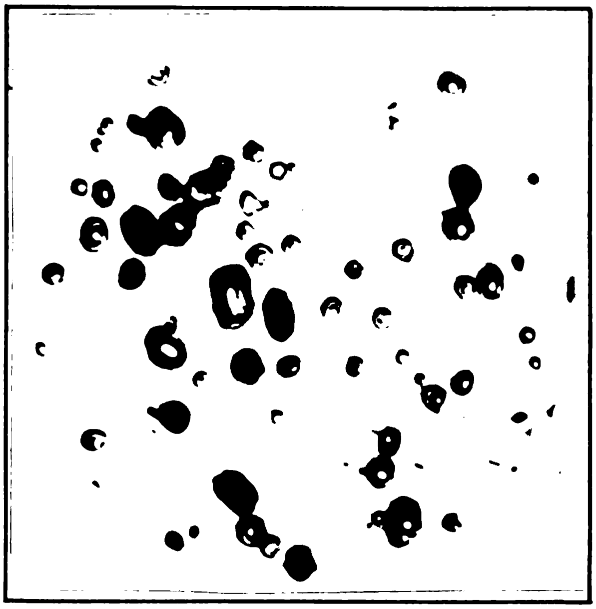
X 2400

Figure 62.

X 2400

Figure 63.

X 2400



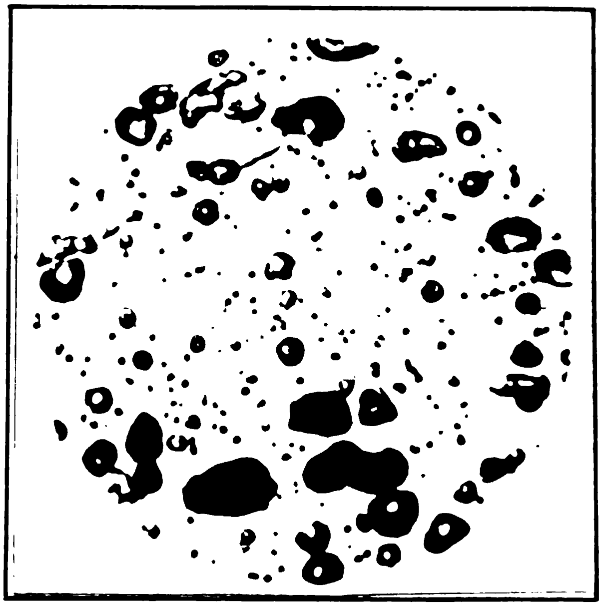


Figure 64.

X 2400

Figure 65.

X 2400

seconds, while the hardened sample was held fifteen seconds. The etching solution was used hot. It shows the tungsten segregation.

Figures 66 and 67 show the annealed and hardened samples respectively after etching with glycerol, nitric acid and hydrochloric acid mixture. The etch shows a very clear and well defined structure. In the hardened sample there is strong evidence of a superimposed structural condition.

Figures 68 and 69 show the annealed and hardened specimens respectively after a very light etch with 4% Nital. The annealed sample was etched ten seconds and the hardened sample four minutes. There is again definite evidence of a superimposed structure in the hardened sample.

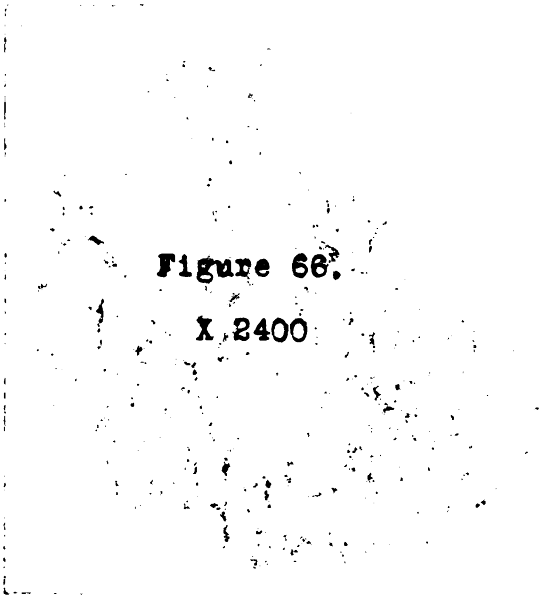


Figure 66.

X 2400

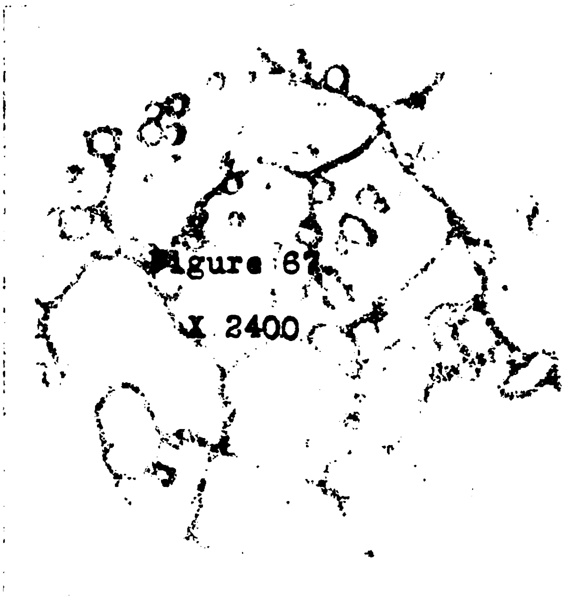
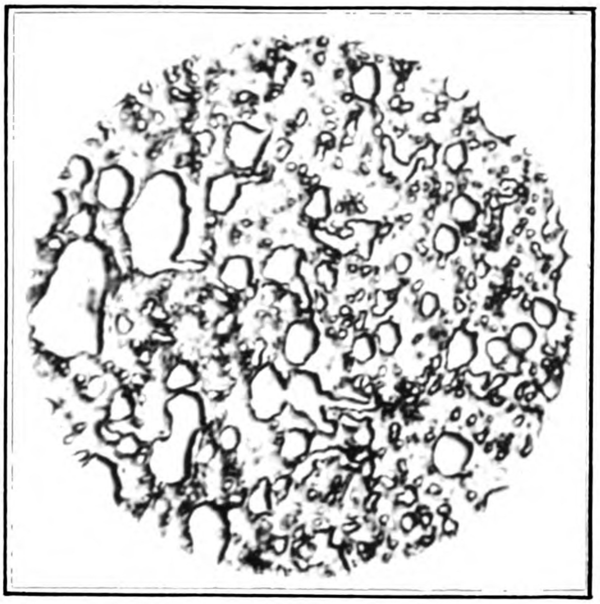
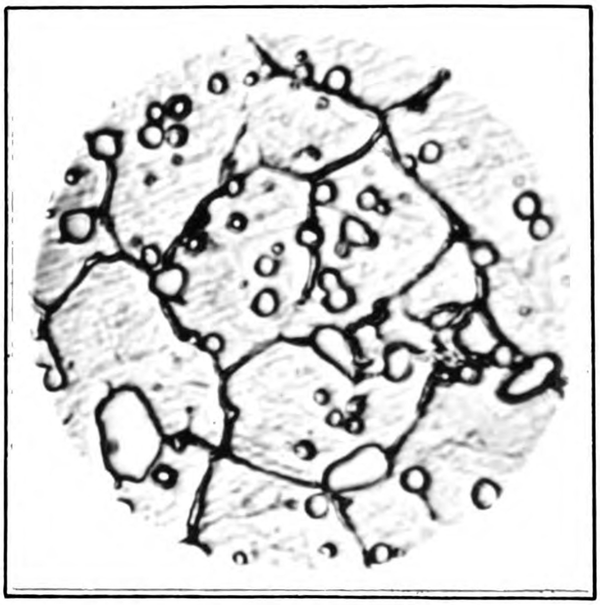


Figure 67

X 2400





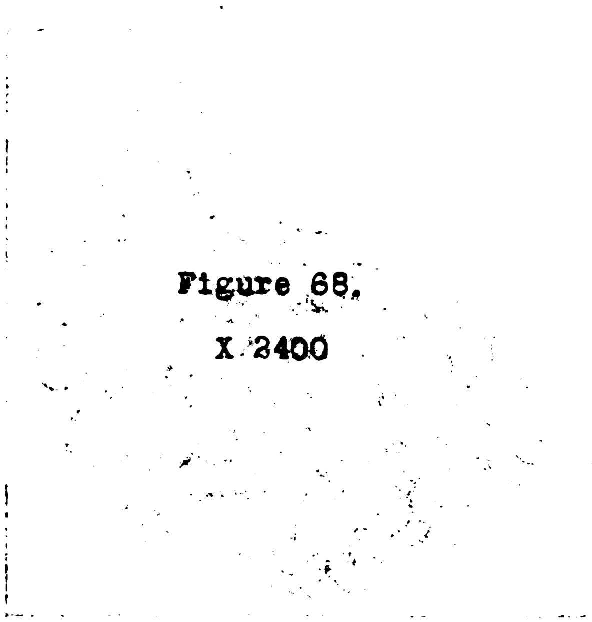


Figure 68.

X 2400

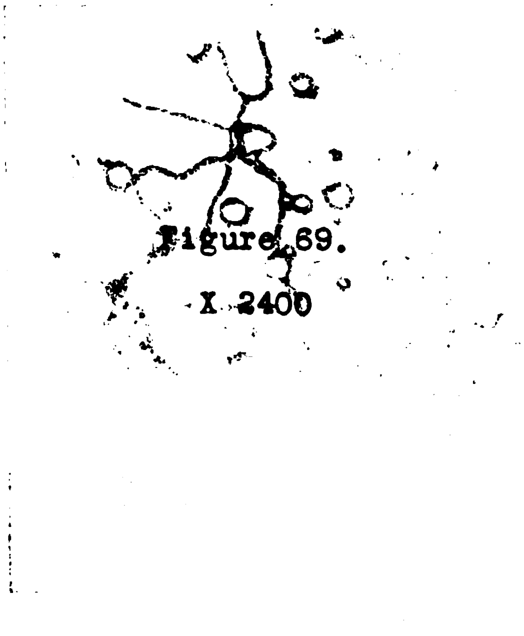
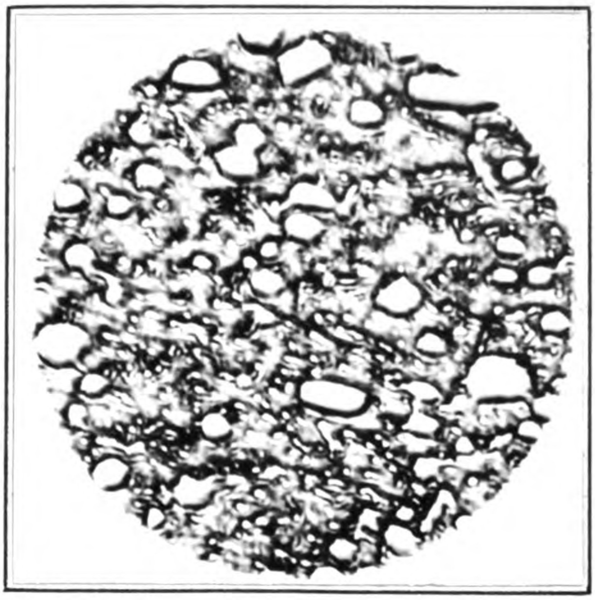
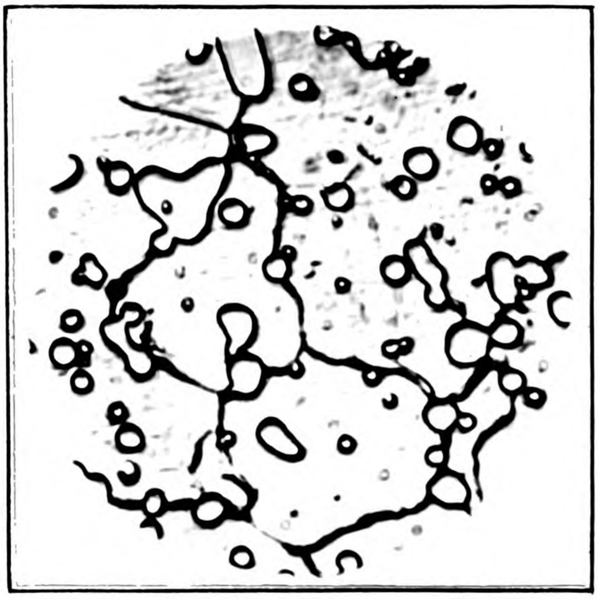


Figure 69.

X 2400





DILATOMETRIC STUDY.

Dilation curves were taken using the Chevenard Industrial Thermal Analyzer*. Figure 70 shows a picture of the instrument, while Figure 71 shows the working principle.

A cylindrical test specimen 2.12 inches long and 0.63 inches in diameter, with a one-fourth inch hole through the center, is placed in the silica tube. The closed end of this tube is square with the horizontal axis. The specimen is so placed that it rests firmly against the closed end of the silica tube. A small silica rod rests against the other end of the specimen and connects through a lever system to a pen arm. Any movement of the sample is magnified about 70 times by the pen. The chart upon which the dilation is recorded is carried on a revolving drum, the speed of which can be regulated. Within the sample is placed a small standard pyros rod. This rod also bears against the closed end of the silica tube, and is connected by a mechanism similar to that of the test specimen, to a pen. The movement of the pen is recorded in degrees of temperature. Thus, we have two curves drawn simultaneously, one showing changes in length of the specimen, the other changes in length of the pyros rod; the latter being calibrated in degrees of temperature. Using the data obtained from the two curves, another graph may be plotted using change in length and temperature as co-ordinates.

* Descriptive matter from; Some Observations on the Dilation of Several Common Steels. Publow, Heath, Batchelor.

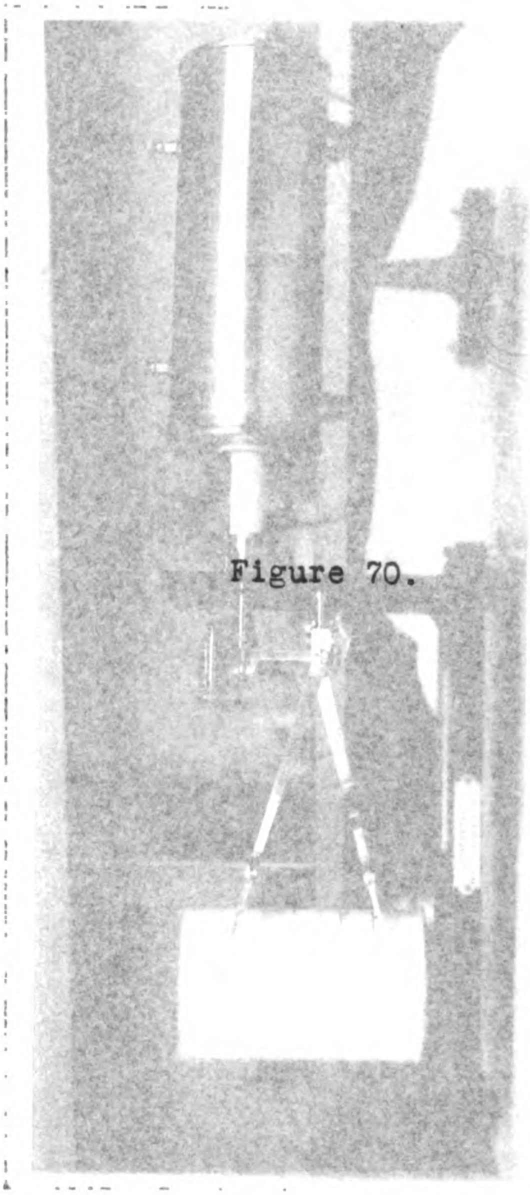


Figure 70.

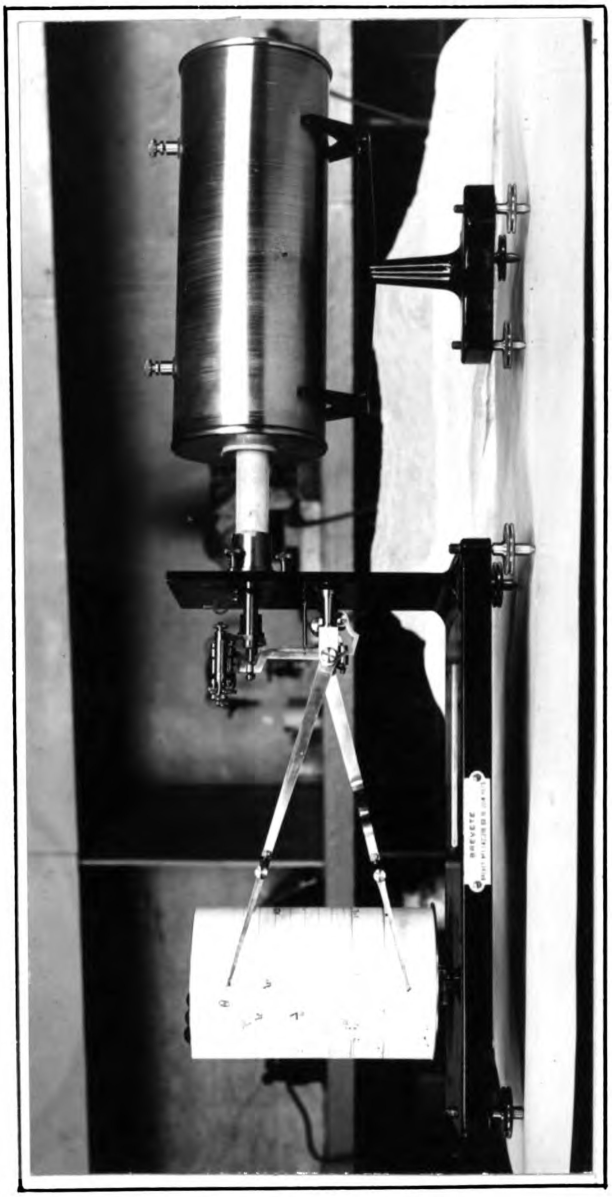
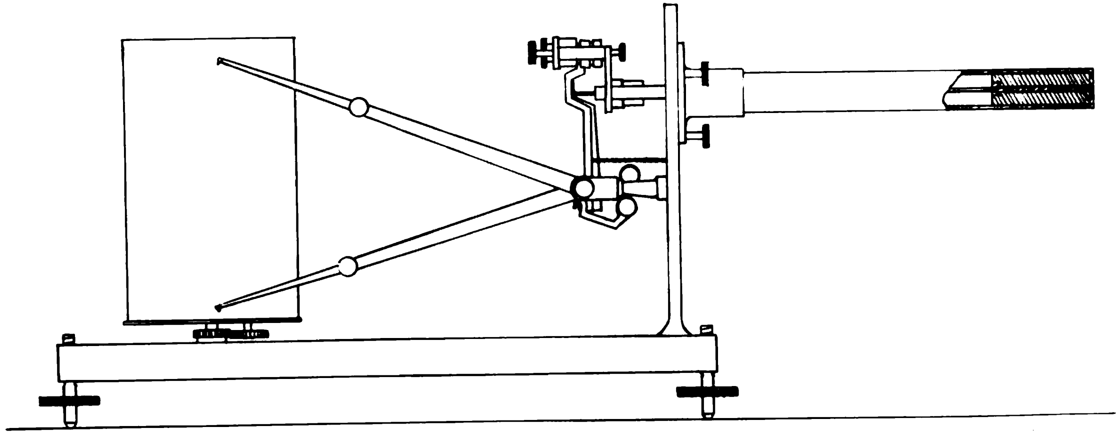

BREVETE

Figure 71.



CHEVENARD **DILATOMETER**

Figure 72 shows a copy of the original two curve chart recorded by the Chevenard Dilatometer.

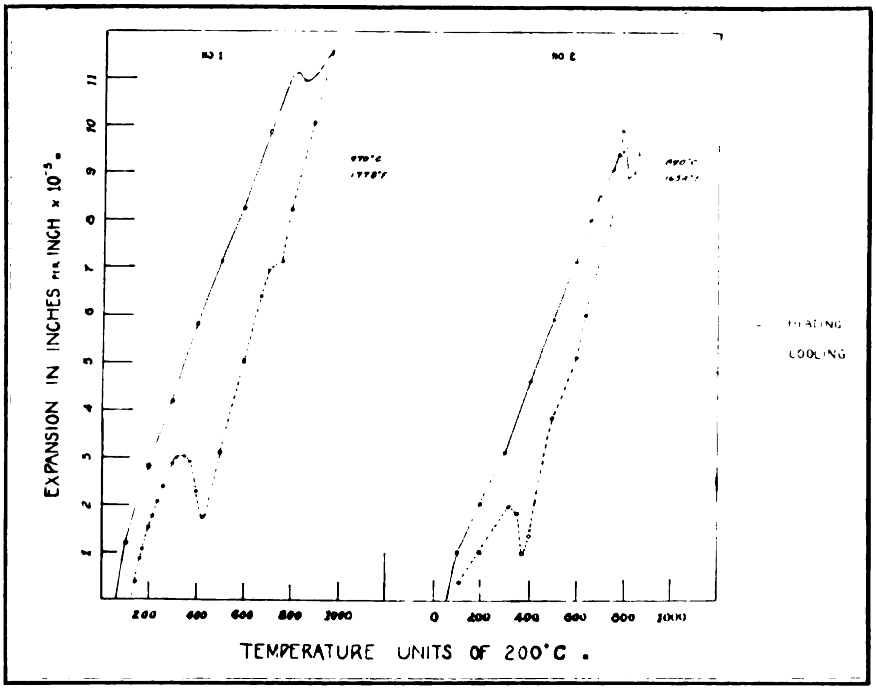
In Figure 73 (No.1) the above two curves have been plotted using temperature and dilation as co-ordinates.

The test specimens were prepared to the standard size.

The test specimens were run in a variety of ways. The two variables, time and temperature, were used freely. Samples were heated slowly and cooled slowly, such as Figure 72 and its corresponding plotted graph No. 1 in Figure 73. The temperature to which specimen was heated was 970° C (1778° F). The uneven line on heating is due to varying the resistance while temperature was rising. The first sign of a transformation appeared at 825° C (1517° F) and appears to be completed a t 900° C (1625° F). The thermal curve, see Figure 72, shows only a very slight absorption of heat, whereas the dilation curve shows considerable shrinkage in the specimen. The time of heating was 72 minutes; cooling was then started. The Ar transformation commenced after 9.6 minutes of cooling, when temperature had reached 725° C (1337° F). The transformation was completed at 675° C (1245° F). The thermal change was of even less magnitude than that on heating. The dilation change was not pronounced. A normal cooling curve was again assumed. At 420° C (788° F) another retardation in cooling on the thermal curve, and a decided expansion on the dilation curve is noticed. This point is spread over a temperature range from 420° C (778°F to 350° C (662° F). The curves again assumed normal cooling rates down to atmospheric temperature.

Figure 72.

Figure 73,


EXPANSION IN INCHES PER INCH × 10⁻⁵.
NO 1
NO 2
1
2
3
4
5
6
7
8
9
10
11
970°C
1778°F
890°C
1634°F
HEATING
COOLING
200
400
600
800
1000
0
200
400
600
800
1000
TEMPERATURE UNITS OF 200°C.

The complete series of dilation and thermal curves, with their corresponding plotted graphs, are shown in Figures 72 to 91. The rate of heating, elevation of temperature and rate of cooling were varied considerably. This data is shown in detail separately.

It appears as though the Ac transformation is quite constant and independent of any variable factor, such as rate of heating. Its range, however, is spread over at least 100°C (212°F) with the same specimen run under similar or varying conditions. It occurs so consistently at 800° C(1472° F.) to 900° C(1652° F) that we may regard this as the transformation range on heating. The average point appears to be approximately 825° C(1517° F). A few tests recorded points below 800° C the lowest was 750° C (1382° F).

It should be mentioned that at a point approximating 700° C (1292° F) on the thermal curve, there is distinct evidence of a slight change of contour. It is just possible that this, rather than being a separate and distinct critical point, is the beginning of the transformation shown at a higher temperature. This steel is naturally sluggish to absorption of heat or any physical change. Therefore, it seems highly possible that the transformation of alpha iron to the solid solution of gamma iron is equally sluggish, and that the transformation shows its first movement as low as

Figure 74.

NO.2

Figure 75.

Figure 76.

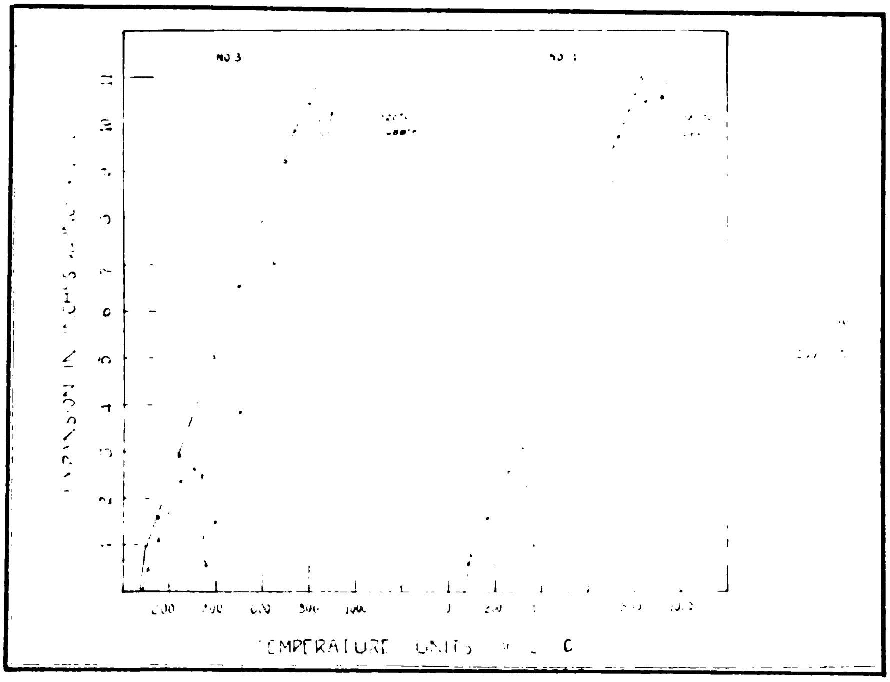

NO 3
11
10
6
5
4
3
2
200
500
TEMPERATURE UNITS
C

Figure 77.

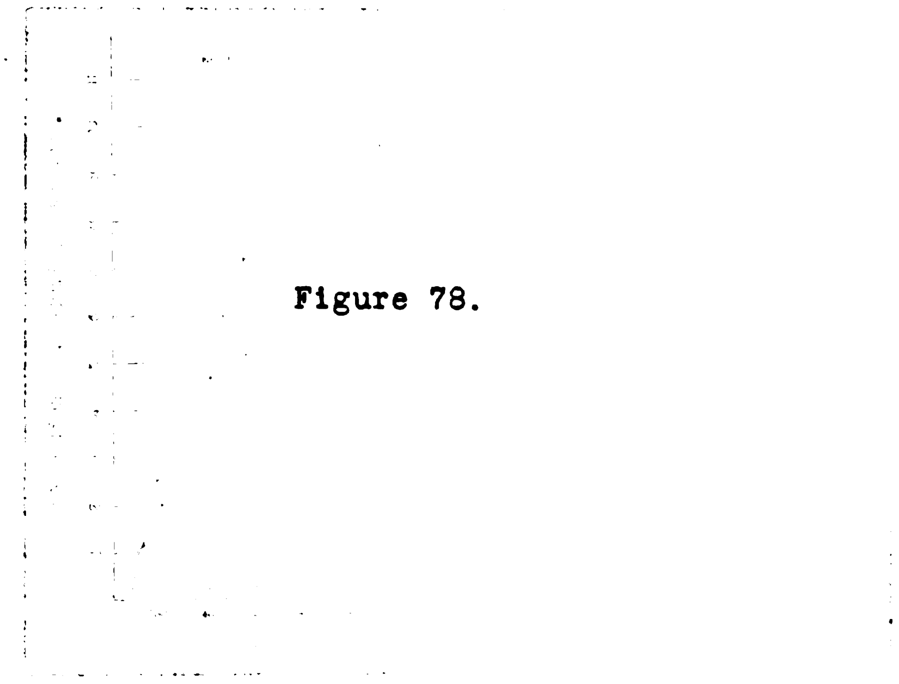


Figure 78.

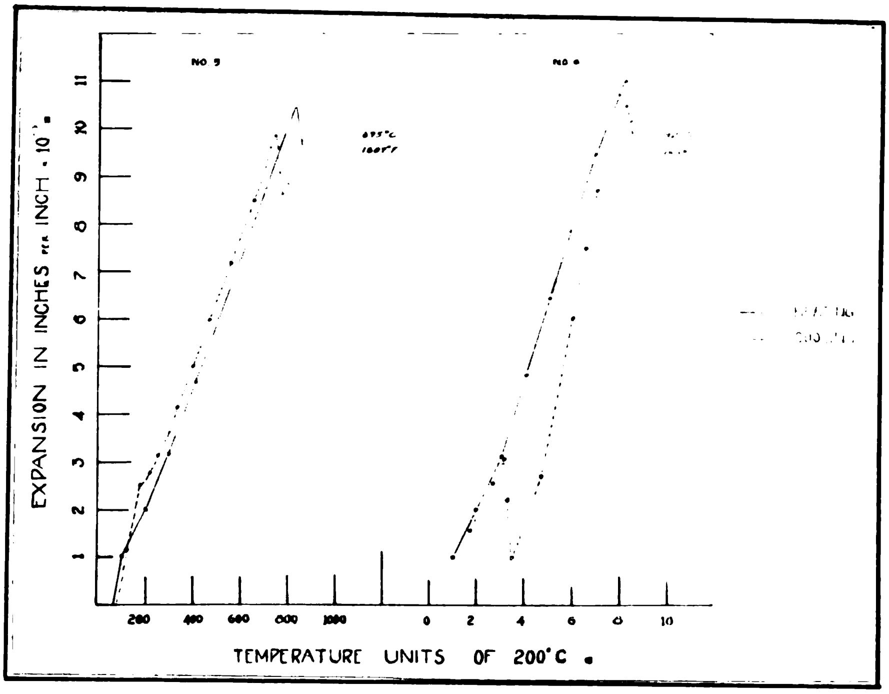
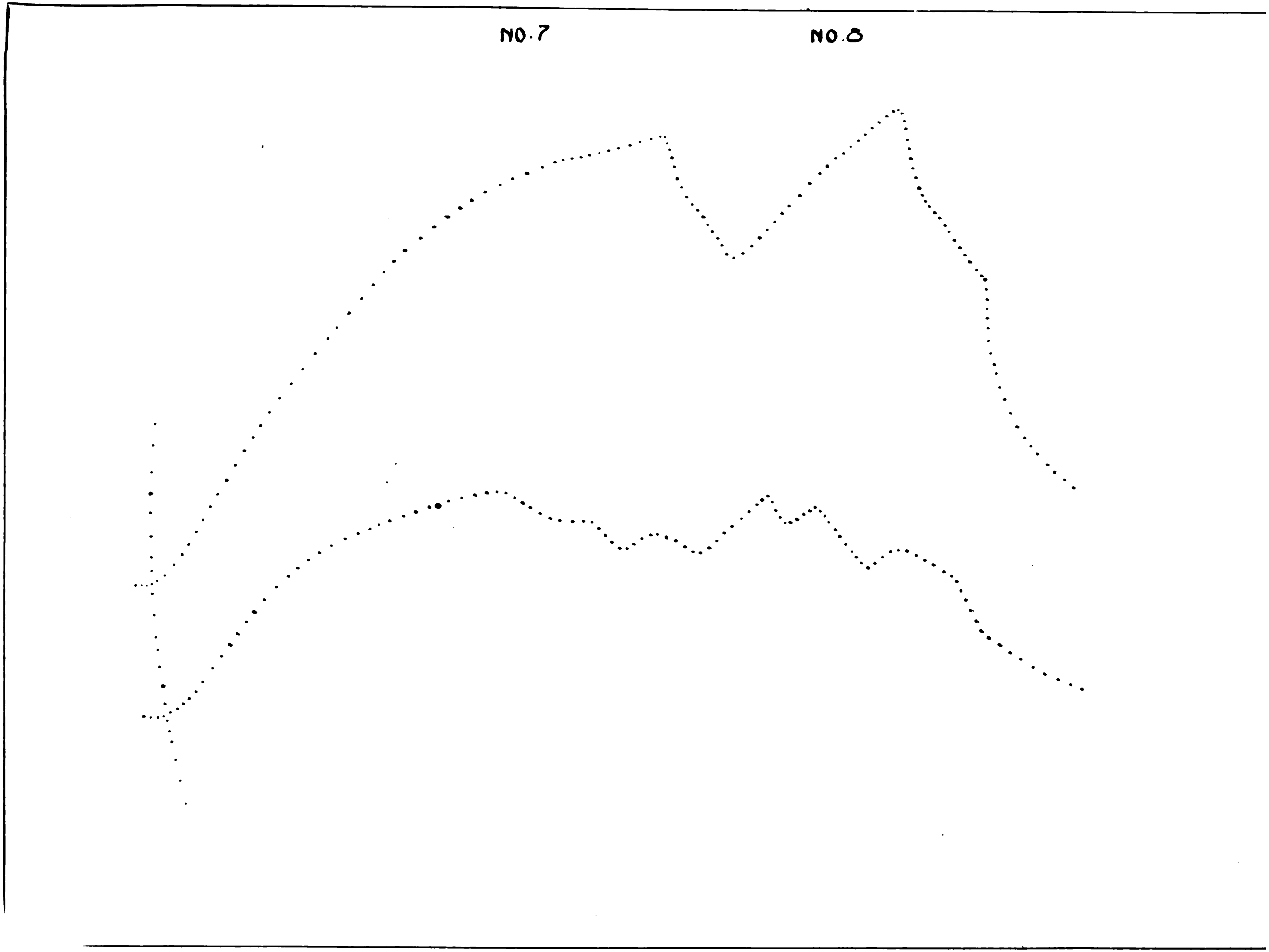

NO. 5
NO. 6
875°C
1607°F
HEATING
COOLING
EXPANSION IN INCHES PER INCH · 10⁻³
1
2
3
4
5
6
7
8
9
10
11
200
400
600
800
1000
0
2
4
6
8
10
TEMPERATURE UNITS OF 200°C

Figure 79.


NO.7
NO.8

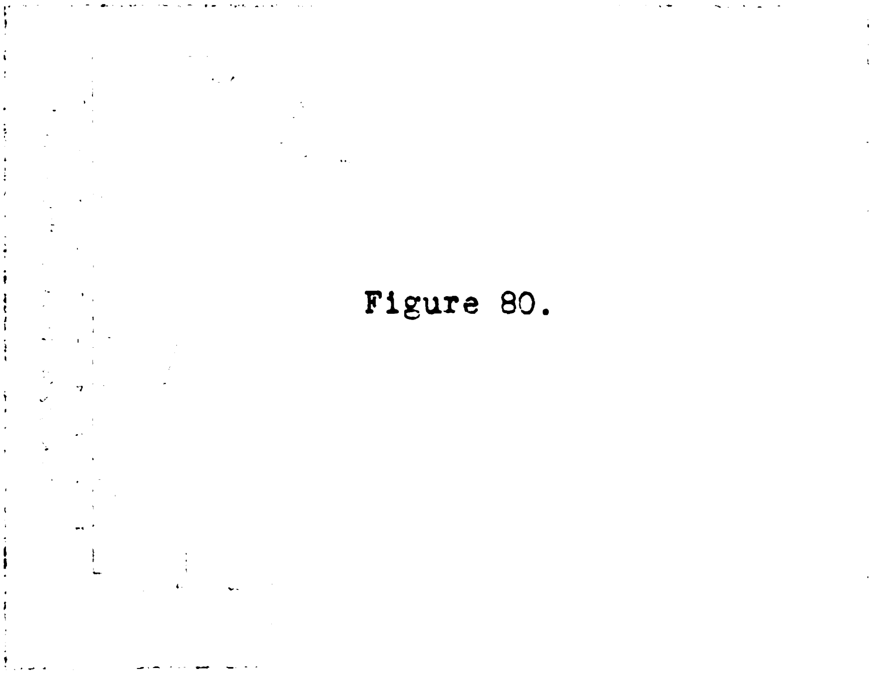
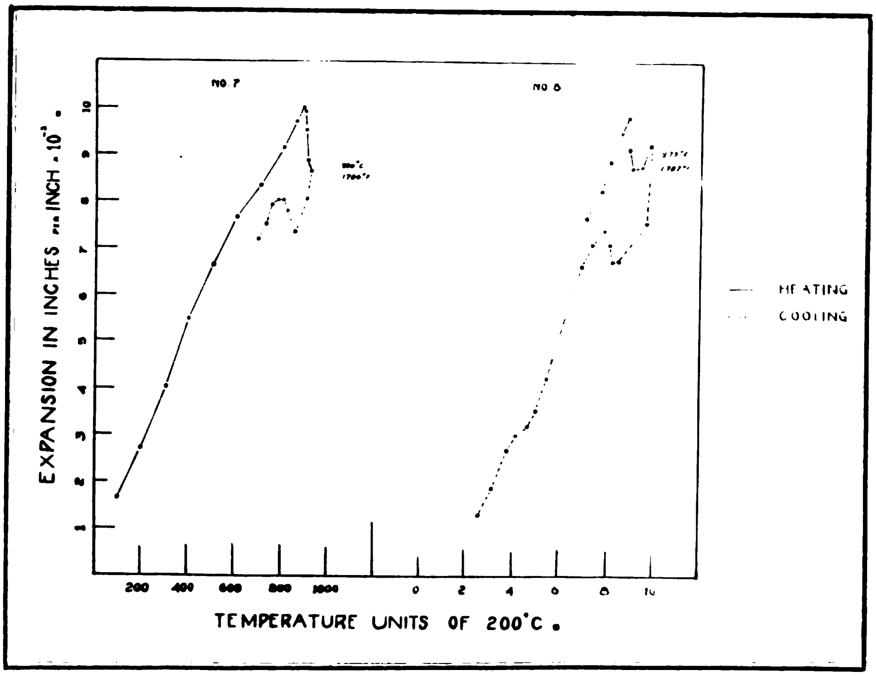


Figure 80.



NO. 7
NO 8
EXPANSION IN INCHES PER INCH · 10^{-3}
1
2
3
4
5
6
7
8
9
10
200
400
600
800
1000
0
2
4
6
8
10
TEMPERATURE UNITS OF 200°C.
HEATING
COOLING

Figure 81.

NO.9

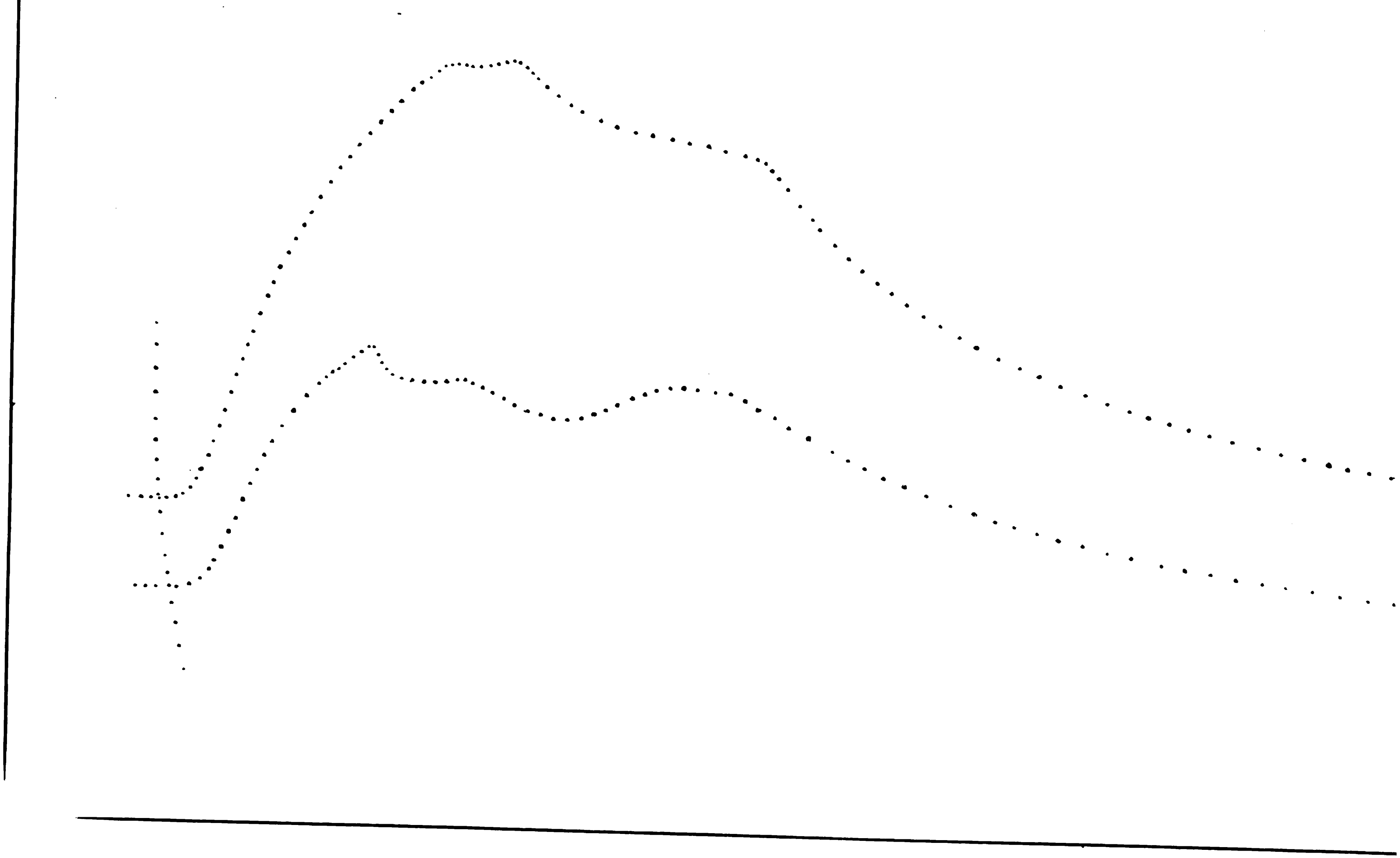


Figure 82.

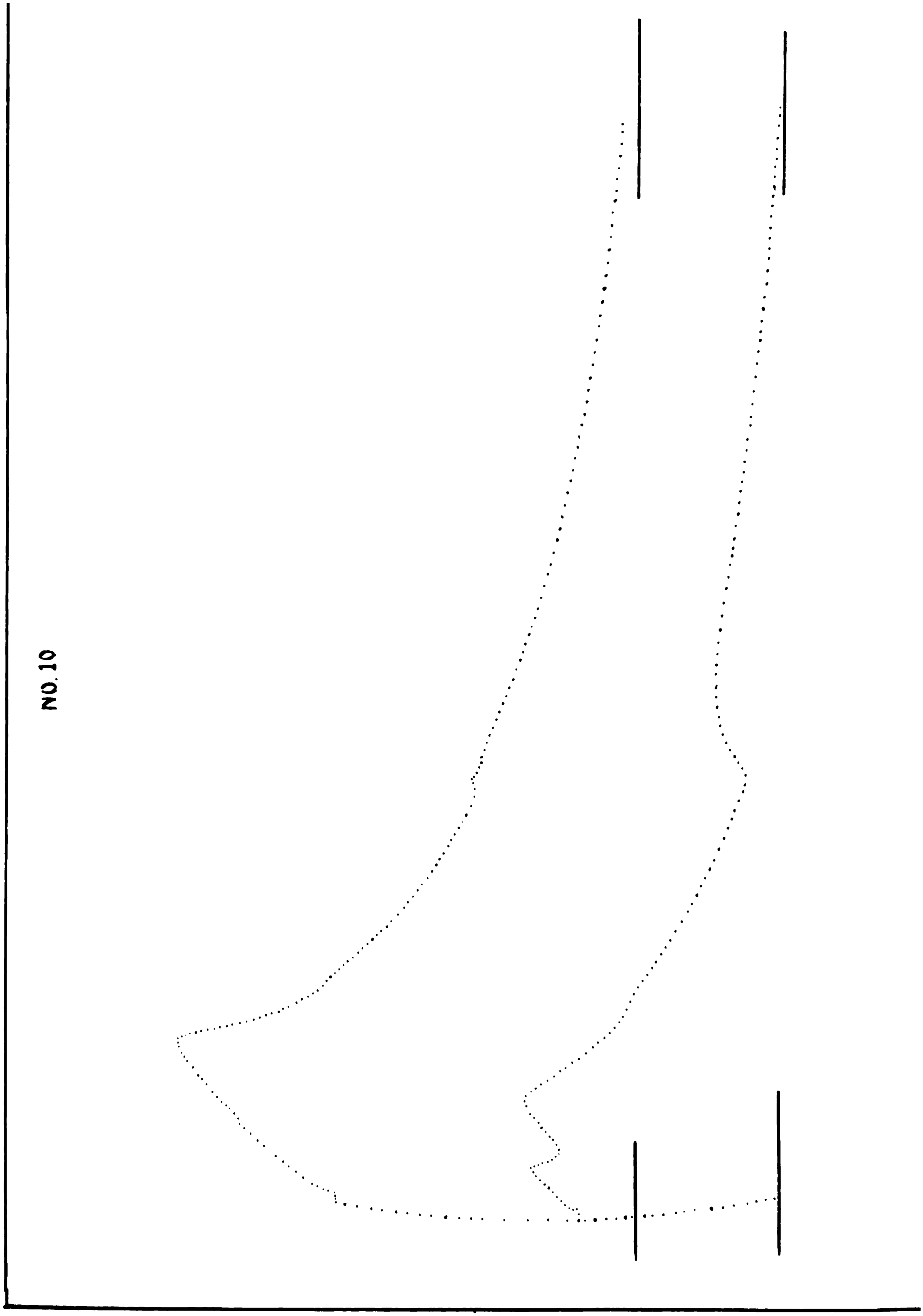
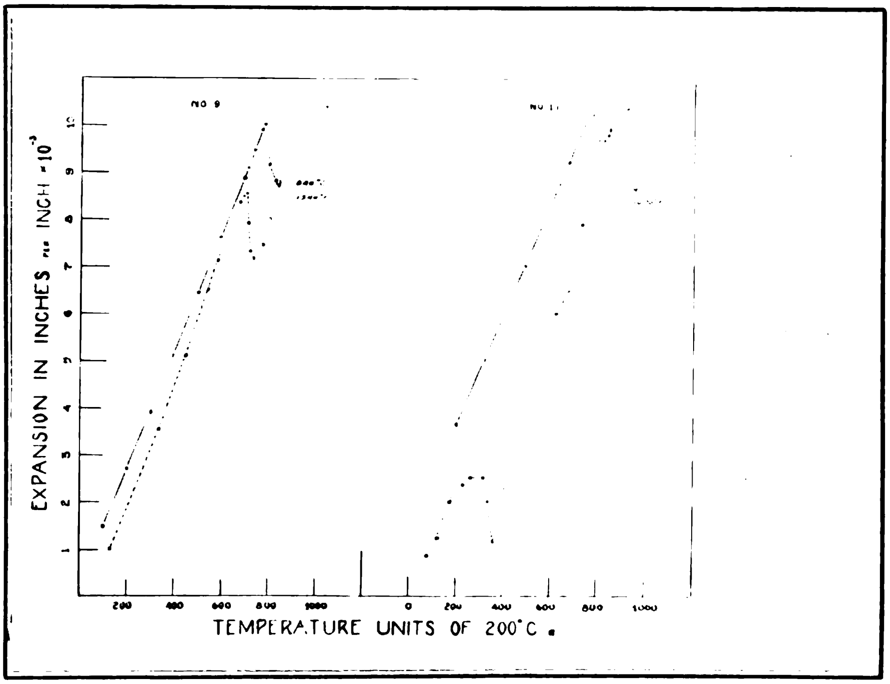

NO.10

Figure 83.


NO 9
NO 11
EXPANSION IN INCHES PER INCH × 10⁻³
1
2
3
4
5
6
7
8
9
10
200
400
600
800
1000
0
200
400
600
800
1000
TEMPERATURE UNITS OF 200°C

700°C(1292°F) and is not completed until 900°C(1652°F) or thereabouts. This thought is somewhat borne out by the gradual hardening, see Figure 41, with increasing temperatures. At 1480°F(1804°C) the Rockwell hardness increased from C 16 to C 45, but below this temperature no hardening took place. Of course, the maximum hardness, when based upon facts from hyper-eutectoid steels, should be obtained immediately the Ac transformation is completed.

The writer feels the maximum hardness resulting from the alpha to gamma iron transformation is probably obtained at 900°C(1652°F),but that the super-hardness of this steel is due in no small measure to increased solution of excess carbides and the heterogeneity of the carbides of iron and tungsten retained throughout the matrix on quenching from a very high temperature. This latter condition might be likened to the interference theory using grain boundaries instead of crystal lattice planes. An alternative thought for this slight thermal change on heating (700°C)is that it could be the beginning of the solution of excess carbides. This would account for the lack of hardening within this narrow range below 800°C(1472°F).

The transformation points on cooling do not show the regularity of the heating point. Depending on elevation of temperature or rate of cooling, we may have one or two Ar critical points.

Figure 84.

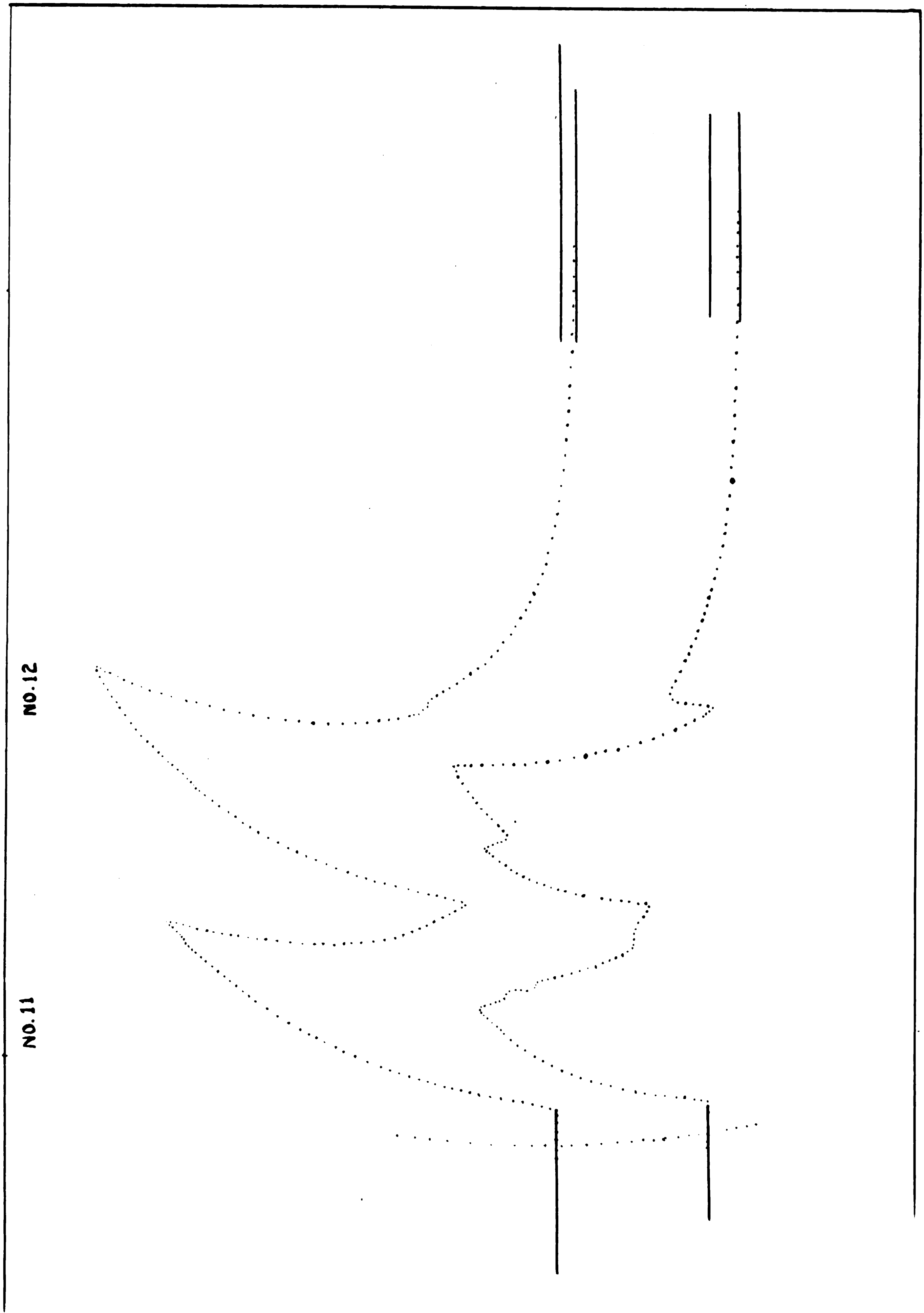

NO.12
NO.11

Figure 85.

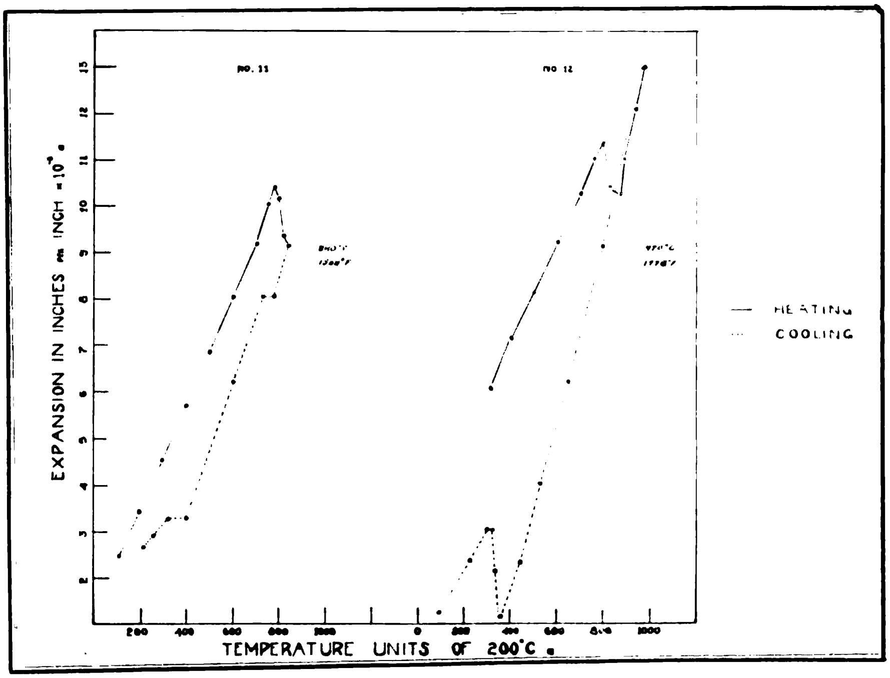

NO. 11
NO 12
EXPANSION IN INCHES PER INCH × 10⁻⁴
2
3
4
5
6
7
8
9
10
11
12
13
200
400
600
800
1000
0
200
400
600
800
1000
TEMPERATURE UNITS OF 200°C
HEATING
COOLING

Figure 86.

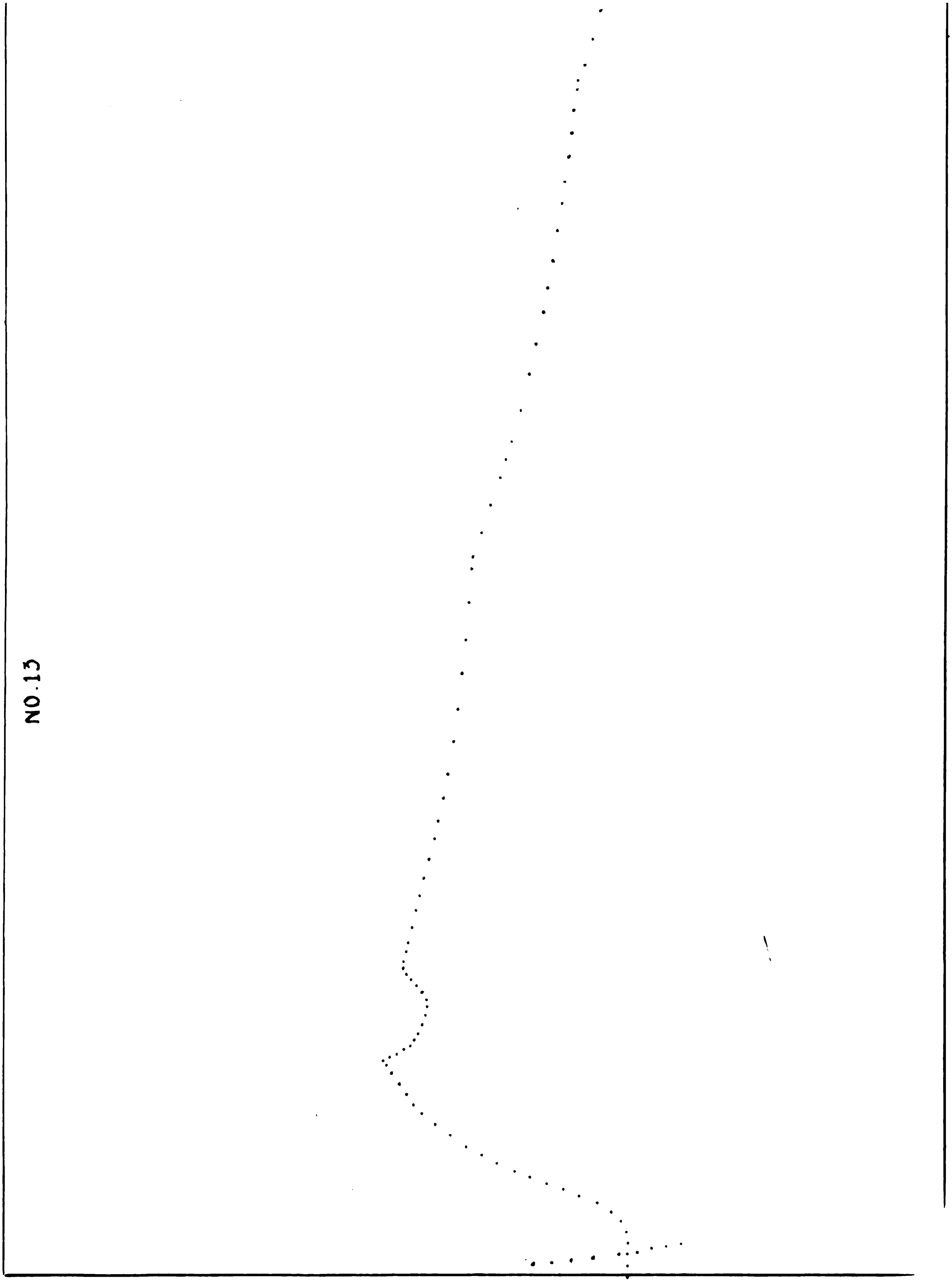

NO.13

Figure 87.

NO. 14

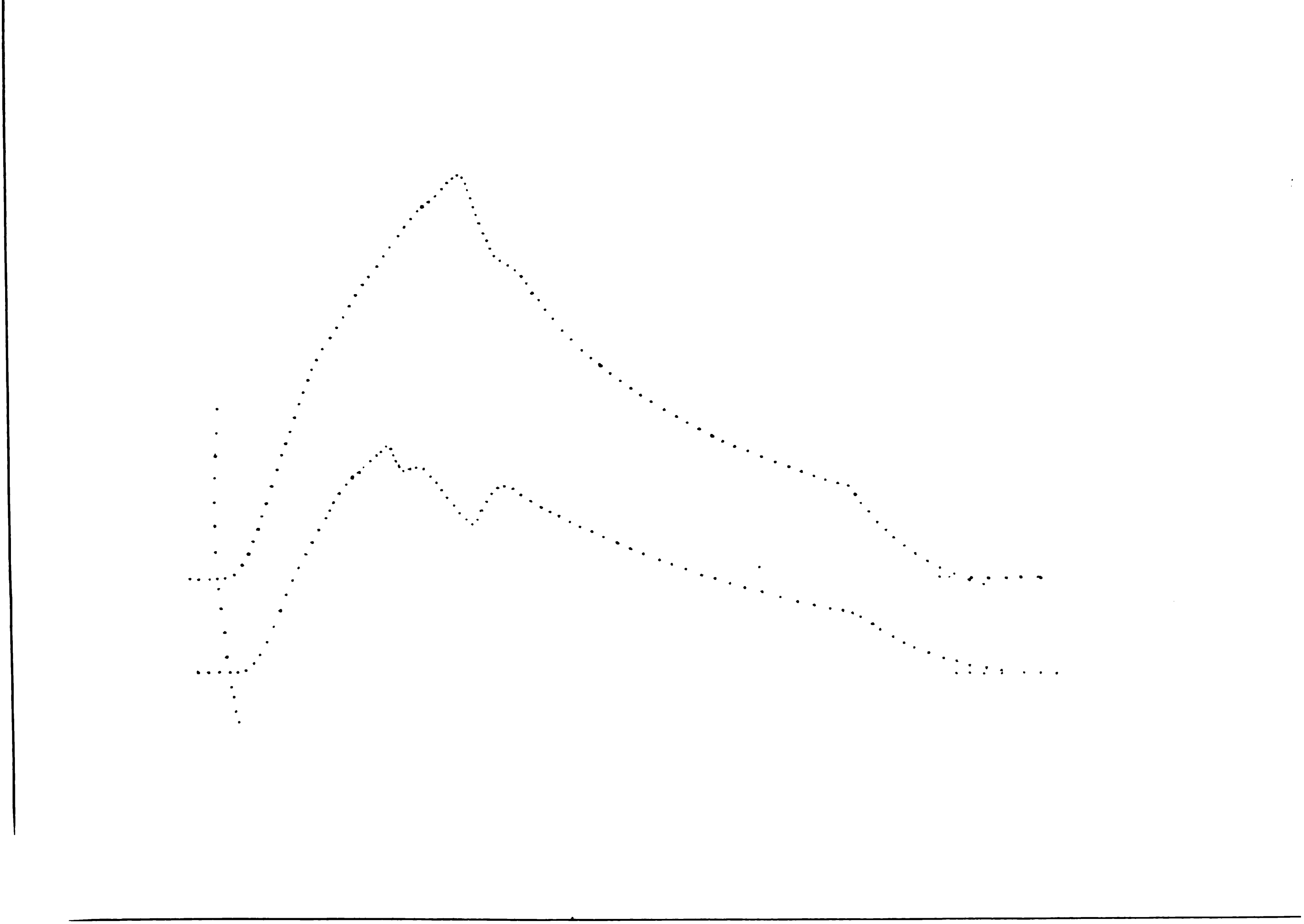
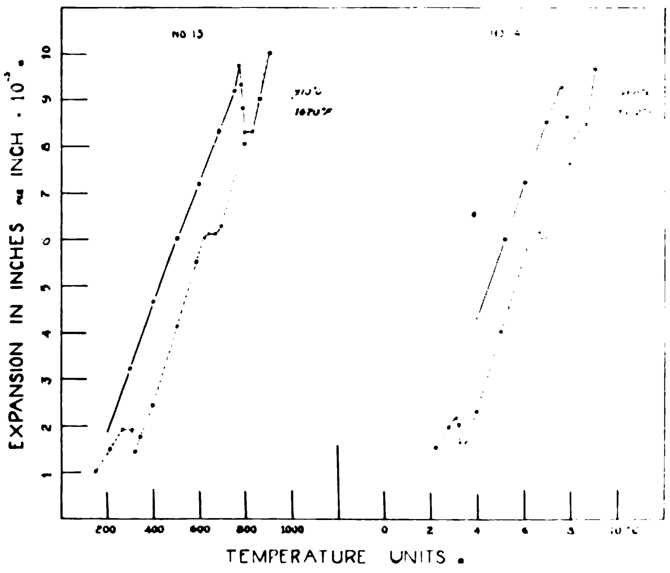


Figure 88.


EXPANSION IN INCHES PER INCH · 10⁻³
1
2
3
4
5
6
7
8
9
10
NO 13
910°C
1670°F
200
400
600
800
1000
0
2
4
6
8
10 °C
TEMPERATURE UNITS

When heated at or above 900°C(1652°F) and cooled slowly, see Figures 83, 84, 87 and 88, two points appear, namely, Ar 700/800°C(1292/1472°F) and Ar$_2$ 265/490°C(510/914°F). An exceptionally wide range of temperature for different samples is involved. The above temperatures express the maximum range for all samples. By reference to individual graphs, the particular point or points can be studied for each test.

We have one exception to the above statement. Figures 85 and 86 show graphs of the sample which was heated to 840°C (1544°F) and cooled at a fast rate, 32$\frac{1}{2}^{\circ}$ per minute. Two cooling points were observed, Ar 690/765°C(1274/1409°F) and Ar$_2$ 315/400°C(600/752°F). It is most difficult to account for this phenomenon when several samples were heated between this temperature and 900°C(1652°F) without showing more than one point on cooling.

The inconsistency of the Ar points leads one to believe that they might be governed to a certain extent by composition (solution), or perhaps precipitation of carbides. The true crystal lattice change is perhaps shadowed by a chemical phenomenon.

The Ar$_2$ dilation point is ordinarily of much greater intensity thanthe Ar$_1$point. This, however, can be altered by use of the variable factors at hand. If the sample is cooled immediately after the Ac transformation is completed, see Figures 78, 79, 82 and 84, the Ar_1 point becomes very pronounced, and the Ar_2 point is completely eliminated. This method of cooling produces an exceptionally low Rockwell hardness, as low as C 4 to C 6, see Figures 78 and 79.

Inasmuch as it is generally regarded that the carbon in this steel exists as a complex carbide of iron and tungsten, is it not reasonable to surmise that considerable time and elevation of temperature is required to break up this constituent and cause solution of the carbon in the austenite? By cooling immediately after the Ac transformation, this necessary time and temperature required for solution of carbon is denied, and the resulting or precipitated phase (austenite-martensite) will be low in carbon and consequently of a soft nature. This is proven to a certain degree by the extremely high quenching temperature necessary to obtain maximum hardness. The low carbon content of the austenite materially raises the temperature of transformation. Three tests, see Figure 72, 85 and 80 were run at 970°C (1778°F),970°C, and 975°C(1787°F) respectively. Figures 72 and 80 were cooled at a moderate rate, and show two well defined points which check with each other very closely.

Figure 85 shows the same sample after a quick cooling. One point of considerable intensity is observed at 290°C(554°F) to 315°C(600°F). This is undoubtedly due to the Ar_1 point being denied time for formation, and the two points are ex-

Figure 89.

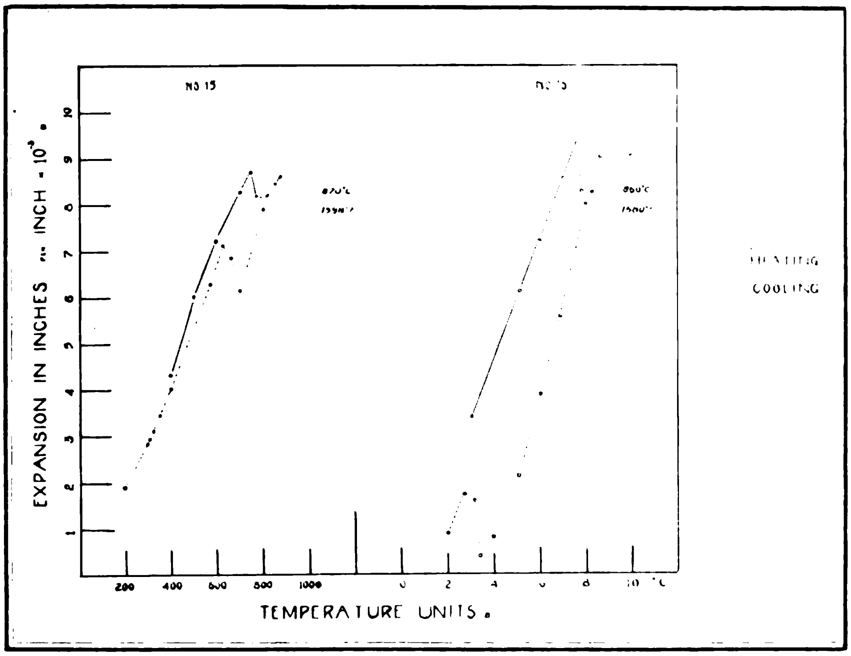
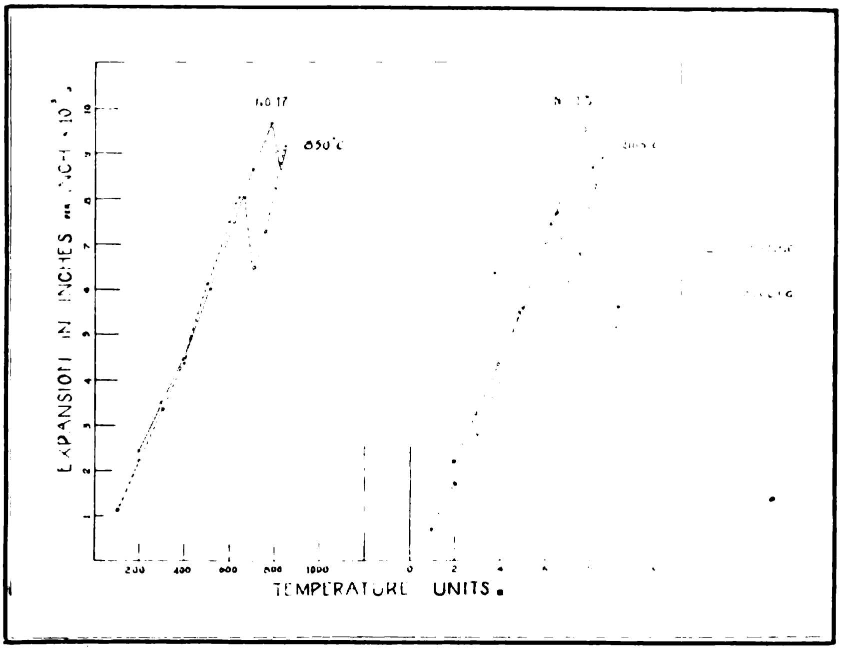

NO 15
870°C
1598°F
860°C
1580°F
HEATING
COOLING
EXPANSION IN INCHES PER INCH × 10⁻³
1
2
3
4
5
6
7
8
9
10
200
400
600
800
1000
0
2
4
6
8
10 °C
TEMPERATURE UNITS

Figure 90.

Figure 91.


EXPANSION IN INCHES PER INCH × 10³
1
2
3
4
5
6
7
8
9
10
No 17
850°C
200
400
600
800
1000
0
2
4
6
TEMPERATURE UNITS

pressed as one. According to results published by Dr. Albert Sauveur he gives two Ac points, one at 775°C(1427°F), and the other at 855°C(1571°F). In the opinion of the writer these should be regarded as the same point, but the sluggishness of the reaction causes a variation of at least 100°C on similar tests.

The writer believes that between 800° and 900°C the transformation from alpha ferrite to a solid solution of gamma (austenite) iron is completed. Microscopic evidence of an austenitic condition after quenching from this point is lacking, but such a condition is difficult to show.

The increased hardness resulting from quenching above this temperature is thought by the writer to be due entirely to solution of carbon and carbides and the heterogeneity of carbides retained in the solid.

The martensite precipitated after a high quench will, of course, have a much higher carbon (hardening) content if time and temperature is given for dissociation of carbides and solution of carbon by the solid solution of austenite. The hardness of martensite is variable due to its carbon content. This carbon percentage is of course due to the percentage of carbon, elemental or as a compound, in solution in the preceeding phase, namely austenite.

After observing the exceptionally low Rockwell hardness obtained by cooling specimen slowly immediately after the Ac transformation was completed, see Figures 79 and 84, it was decided to repeat this test with hardened samples quickly and slowly cooled immediately after completion of the gamma to alpha iron (Ar) transformation. These tests are shown in Figures 88, 90 and 91. The first sample run, graph shown in Figure 88, had a hardness of C59/60 after hardening at 2100°F(1149°C) in air. Specimen was heated to 820°C(1510°F) and cooled slowly until the Ar transformation was completed and then allowed to cool slowly in tube of furnace. Rate of cooling was 26.6 deg.Cent. per minute to the Ar point and from there to atmospheric temperature was 13.8°C per minute. The Rockwell hardness resulting from this treatment was C10/15. This curve was not plotted as it is not materially different from many of the preceeding graphs.

Figs.90 (17 and 18) and 91 show the dilatometric curve and plotted graphs of the rehardened specimen. The Rockwell hardness after treating at 2050°F(1120°C) in air was C53/55. Both tests were permitted to cool slowly in the silica tube until the Ar transformation was completed, this point was at 635°C(1175°F) in number 17 and 645°C(1193°F) in number 18. The rate of cooling to this point was 23°C and 23½°C per minute. At this temperature the furnace was removed from the containing tube and cooling hastened by fanning. The rate of cooling was 59°C and 57.2°C per minute. The Rockwell hardness resulting was C40/42 and C24/25 respectively. With these two samples, treated identically alike, there is an appreciable variation in Rockwell hardness.

It is to be noted that quick cooling after the Ar transformation has been completed has caused a considerable increase in hardness. Inasmuch as there is no evidence of a critical transformation it is difficult to understand the resulting increase in hardness. Owing to the differential in the increase in hardness resulting from identical treatments it is the thought of the writer that this might be due in a degree to precipitation of carbides and tungstides.

Figure 92 shows a photomicrograph of a section of the test specimen after a slow cool immediately following the Ar transformation. Rockwell hardness C8/10.

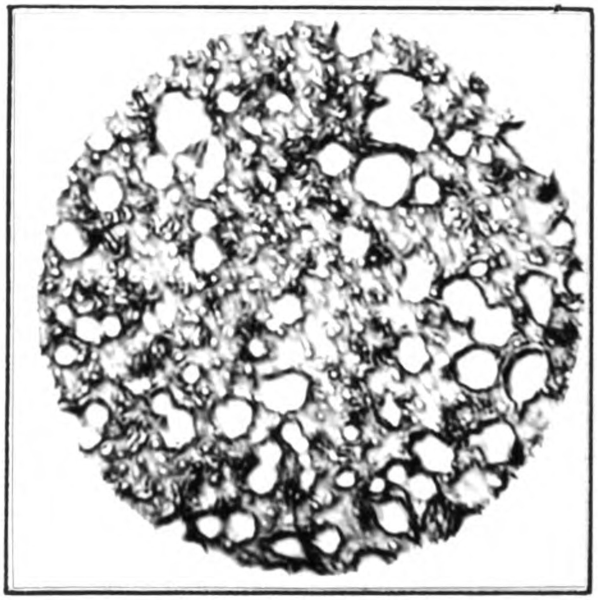
Dilatometric Details

Figure Number	Rate of Heating	Max. Temp.	Ac Point	Rate of Cooling	Ar Point	Rockwell Hardness
1, 72,73.	13°/min.	970°C	825/900°C	25°/min	Ar_1 725/700°C Ar_2 420/380°C	C 35/36
8, 79,80.	19°/min.	975°C	875/900°C	22°/min	Ar_1 820/770°C Ar_2 490/420°C	C 20/24
12, 84,85.	43°/min.	970°C	770/815°C	70°/min	315/290°C	C 50/52
13, 86.	18°/min.	910°C	760/800°C	26°/min	Ar_1 680/630°C Ar_2 320/265°C	----
14, 87.	42°/min.	900°C	755/810°C	26°/min	Ar_1 680/625°C Ar_2 350/310°C	-----
3, 75,76.	22°/min.	920°C	825/850°C	86°/min.	370/360°C	C 39/41
4, 75,76.	41°/min.	920°C	815/850°C	76°/min	375/365°C	C 39/41
7, 79,80.	18°/min	930°C	875/900°C	17°/min	850/800°C	-----

Figure Number	Rate of Heating	Max. Temp.	Ac Point	Rate of Cooling	Ar Point	Rockwell Hardness
10, 82,83.	19°/min	910°C	800/815°C	16°/min	Ar_1 680/630°C Ar_2 360/325°C	C 38/40
2, 73,74.	17°/min	890°C	800/825°C	14°/min	380/320°C	C 39/41
5, 77,78.	14°/min	875°C	830/860°C	7°/min	780/740°C	C 4/6
9, 81,83.	26°/min	840°C	790/830°C	10°/min	745/715°C	C 8/11
11, 84,85.	50°/min	840°C	780/815°C	53°/min	Ar_1 765/690°C Ar_2 400/315°C	-----
15, 89.	42°/min	870°C	750/820°C	26°/min	700/625°C	-----
16, 89.	40°/min	860°C	760/820°C	102°/min	340/270°C	C 28/29

Figure 92.

X 2400



CONCLUSIONS

1. The hardening of high speed steel varies with the temperature of quench, reaching a maximum at a point just preceeding incipient fusion. Going through this point causes excessive grain growth with a softening and embrittling effect. If held for a long period of time complete diffusion of carbon takes place.

2. After passing through the critical range on heating, 800° to 900° C, the increased hardness is due to the increased solution of carbon or carbides resulting from time and temperature given sample. The heterogeneity of precipitated carbides is shown in photomicrographs and without doubt is a large factor in the hardness and stability of this steel at elevated temperatures.

3. Elevation of temperature has little or no effect in breaking up the carbides and eutectics formed at time of pouring ingot.

4. The irregularity of the Ar critical points of this steel when cooled from varying temperatures is due entirely to the varying percentages of carbon going into solution with the gamma iron, and the corresponding varying percentage in the martensite formed on cooling through this critical range. The hardness of this latter structure varies directly with the elevation of temperature to which the specimen was carried.

5. Martensite is of variable carbon composition.

6. Rate of cooling high speed steel after passing through the Ar_1 point increases the hardness slightly. This applies only to steels cooled immediately after completion of the Ac transformation and consequently having austenite of low carbon content.